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VOL. 164 PART I

Principal Contents

Overheating and Burning of Steel—Part III
by A. Preece, J. Nutting and A. Hartley

Detection of Overheating and Burning in Steel by Microscopical Methods... *by A. Preece and J. Nutting*

Grain-Boundary Phenomena in Severely Heated Steel
by T. Ko and D. Hanson

Tessellated Stresses—Part V ... *by F. Laszlo*

Temper Brittleness ... *by L. D. Jaffe*

Breakdown of Austenite Below the M_s Temperature
by F. C. Thompson and M. D. Jepson

Ingot Heat Conservation
by A. V. Brancker, J. Stringer, and L. H. W. Savage

Historical Note No. 15 ... *by H. R. Schubert*

Split Thermocouples ... *by A. Searby*

Wire-Drawing-Die Profilometer ... *by R. M. J. Withers*

Discussion on Spent Acid Pickling Liquor, and Annealing Plant for Mild Steel Sheets and Coils

News and Abstracts

A full list of Contents is given on the Title Page (see page 41, Advertisement Section)

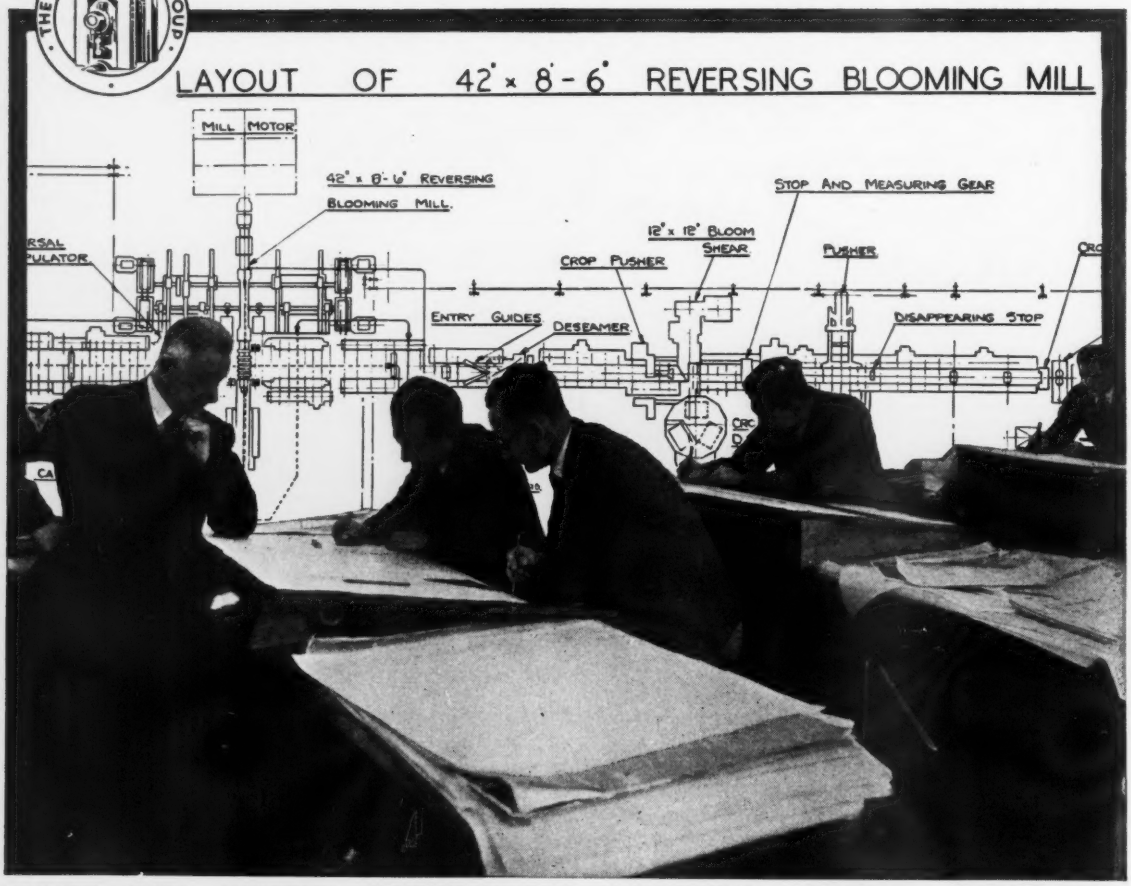
PRODUCTION STORY — LAYOUT

ROLLING MILL PLANT does not conform easily to set or preconceived designs. Each mill has features peculiar to itself, dependent upon the productive requirements to be met.

Given a choice the Layout Engineers would naturally plump for the " green field " of an entirely new metal works plant where, as often before, they can design the whole plant as a complete entity. Sometimes, however, they have to fit a new mill into a works layout planned around very different equipment.

A case in point was the recent replacement of a Davy-United 36" Blooming Mill with a 42" mill which had to be framed in the original foundation space. Since no alternative blooming facilities existed in this steelworks the prior concern was to reduce the shut-down period of the plant to a minimum while dismantling the old mill and erecting the new.

When this mill rolled its first ingot the shut-down time had not exceeded 30 days, for which the original planning on the layout drawing board was greatly responsible. It was one of many such jobs familiar to Davy-United Layout Engineers; but who can blame them if they rather prefer the challenge of the provocative " green field " ?



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CONTENTS

VOLUME 164 **1950**

PROCEEDINGS OF THE INSTITUTE

PAGE

Special Summer Meeting in Norway, 1949 161

Report of Council and Statement of Accounts for 1949 385

DISCUSSION AND CORRESPONDENCE ON PAPERS

"The Mechanism of the Freezing of Horizontal Steel Castings," by the Melting and Metallurgical Committee 91

"Some Notes on Brittleness in Mild Steel," by J. Görrissen 163

"Experiences in the Study of Isothermal Transformations," by T. F. Russell and C. Mavrocordatos ... 165

"The Acceleration of the Rate of Isothermal Transformation of Austenite," by M. D. Jepson and F. C. Thompson 165

"Determination of an Isothermal Transformation Diagram with an Optical Dilatometer," by J. K. L. Andersen 165

"The Manufacture of Electric Steel in Great Britain," by the Electric Process Sub-Committee ... 167

"The Söderberg Electrode System," by H. Christiansen, jun., and B. Ydstie 167

"Features from the Planning of an Electric Smelting Plant for Pig Iron at Mo i Rana," by B. M. Müller 170

"The Properties of Olivine and Its Uses for Refractories and Moulding Sands," by K. J. Stenvik ... 171

"Electron Microscope Study of Quenched and Tempered Steel," by J. Trotter and D. McLean ... 288

IRON AND STEEL INSTITUTE PAPERS

PAGE

"Carbide Distribution as a Possible Cause of Temper Brittleness," by L. D. Jaffe 1

Historical Note No. 15—"The Pinners and Wire-Drawers of York, 1296–1762," by H. R. Schubert ... 3

"Tessellated Stresses—Part V," by F. Laszlo 5

"The Influence of Gas/Solid Temperature Differences on Blast-Furnace Operation," by J. Taylor ... 129

"Electrochemical Studies of Protective Coatings on Metals: Pt. II—Resistance and Capacitance Measurements on Painted Steel Immersed in Sea Water," by F. Wormwell and D. M. Brasher ... 141

"Vanadium and Its Alloys: A Review of Previous Work," by W. B. Pearson 149

Historical Note No. 16—"The English Knife in the Early Middle Ages," by H. R. Schubert 160

"The Diffusion of Carbon and the Carburization Process," by J. Taylor 257

"A Comparison of Six Spring Steels," by A. S. Kenneford and G. C. Ellis:

Pt. I—"Hardenability and Resistance to Grain Growth and Decarburization" 265

Pt. II—"Jominy Hardenability and Mechanical Properties after Hardening and Tempering" ... 271

Historical Note No. 17—"The Economic Aspect of Sir Henry Sidney's Steelworks at Robertsbridge, in Sussex, and Boxhurst, in Kent," by H. R. Schubert 278

Historical Note No. 18—"The First Iron Handguns in England," by H. R. Schubert 280

"Treatment of Waste Waters from the Pickling of Steel," by G. E. Eden and G. A. Truesdale ... 281

"The Determination of the Texture of Rolled Sheet from X-Ray Diffraction Photographs," by A. E. DeBarr and B. Roberts 285

"The Extension of the Ar_1 Range with Carbide Formation in Mild Steel due to High-Temperature Treatment," by J. H. Whiteley 399

Historical Note No. 19—"An Early Attempt at Trade-Unionism by Ironworkers in North Derbyshire, in 1814," by H. R. Schubert 405

Historical Note No. 20—"The Earliest Use of British Cast Iron in Machine Construction," by H. R. Schubert 406

"The Influence of Low Percentages of Certain Elements on the Microstructure of Pure Iron–Carbon Alloys and Cast Irons," by W. J. Williams 407

"A Note on the Overheating of Steel," by E. C. Rollason and D. F. T. Roberts 422

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

"The Breakdown of Austenite Below the M_s Temperature," by F. C. Thompson and M. D. Jepson ... 27

"Split Thermocouples," by A. Searby 36

"The Overheating and Burning of Steel: Pt. III—The Influence of Excessive Reheating Temperatures on the Mechanical Properties and the Structure of Alloy Steels," by A. Preece, J. Nutting, and A. Hartley 37

"The Detection of Overheating and Burning in Steel by Microscopical Methods," by A. Preece and J. Nutting 46

"Grain-Boundary Phenomena in Severely Heated Steel," by T. Ko and D. Hanson 51

"A Wire-Drawing-Die Profilometer," by R. M. J. Withers 63

"Ingot Heat Conservation: Time Studies from Casting to Rolling," by A. V. Brancker, J. Stringer, and L. H. W. Savage 67

PAGE

"Proposals for the Modification of a Blast-Furnace Top to Give Controlled Burden Distribution," by E. L. Diamond 173

"The Mechanism of Solidification of Horizontal Steel Castings," by D. R. F. West 182

"Structural Frames for Melting Shops," by J. S. Terrington 195

"Cementiferous Paints: Pt. II—Zinc Phosphate Cements," by J. E. O. Mayne 289

"Blast-Furnace Gas Cleaning: Methods for Calculating the Motions of Particles in a Gas," by J. Stringer 294

"Blast-Furnace Gas Cleaning: An Analysis of Plant Performance," by R. F. Jennings 305

Committee Organization Chart 326

"The Traffic of Iron and Steel Works: The Application of V.H.F. Radio Communication," by H. H. Mardon, H. A. Chapman, and M. D. Brisby 425

"A Proposed Method of Specifying Travel-Motion Performance for Steelworks Overhead Cranes," by D. A. Wise and L. N. Bramley 439

"Characteristics of Scrap in Relation to Bulk Handling," by E. L. Diamond 452

IRON AND STEEL WORKS ENGINEERING

Report of the Tenth Meeting (Afternoon Session)—

General Discussion—"The Disposal of Spent Acid Pickling Liquor," by W. B. Wragge 85

General Discussion—"The Design and Operation of Annealing Plant for Mild-Steel Sheets and Coils," by R. D. Pollard, H. Edwards, and J. B. Davies 87

"Electric Generation and Distribution in Iron and Steel Works," by T. Coxon 327

"Heavy Forging Manipulators," by A. V. Flinn 463

News 92, 223, 349, 475

Translations 96, 226, 354, 480

Abstracts 102, 227, 355, 481

Book Notices 125, 253, 383, 510

New Publications 128, 256, 384, 512

List of Periodicals examined by the Abstracting Service 97

ABBREVIATIONS AND SYMBOLS

Å.	Ångstrom unit(s) = 1×10^{-10} m.
A.C.	air-cooled ; alternating current
A.H.	air-hardened
amp.	ampère(s)
amp.hr.	ampère-hour(s)
approx.	approximately
at.-%	atomic per cent
at.wt.	atomic weight
atm.	atmosphere(s) (pressure)
A.W.G.	American wire-gauge
° Bé.	Baumé (degree)
b.h.p.	brake horse-power
B. & S.	Brown and Sharpe (gauge)
B.o.T.	Board of Trade
b.p.	boiling point
B.T.U.	Board of Trade unit(s)
B.Th.U.	British thermal unit(s)
B.W.G.	Birmingham wire-gauge
° C.	centigrade ; Celsius (degree)
cal.	calorie(s)
c.c.	cubic centimetre(s)
c.d.	current density
c.g.s.	centimetre-gramme-second unit(s)
cm.	centimetre(s)
coeff.	coefficient(s)
conc.	concentrated
const.	constant(s)
cu.	cubic
cwt.	hundredweight(s)
D.C.	direct current
dia.	diameter
dil.	dilute
dm.	decimetre(s)
e.m.f.	electromotive force
e.v.	electron volt(s)
° F.	Fahrenheit (degree)
ft.	foot, feet
ft.lb.	foot-pound(s)
g.	gramme(s)
gal.	gallon(s)
H.F.	high-frequency
h.p.	horse-power
h.p.hr.	horse-power-hour(s)
hr.	hour(s)
in.	inch(es)
in.lb.	inch-pound(s)
I.S.W.G.	Imperial standard wire-gauge
K.	absolute temperature (Kelvin scale)
kg.	kilogramme(s)
kg.cal.	kilogramme-calorie(s)
kg.m.	kilogramme-metre(s)
km.	kilometre(s)
kV.	kilovolt(s)
kVA.	kilovolt-ampère(s)
kW.	kilowatt(s)
kWh.	kilowatt-hour(s)
kX.	crystal Ångstrom(s) = 1000 Siegbahn X-units

lb.	pound(s)
L.F.	low-frequency
M	molar (solution)
m.	metre(s)
m.amp.	milliampère(s)
max.	maximum
mg.	milligramme(s)
min.	minimum ; minute(s)
ml.	millilitre(s)
mm.	millimetre(s)
m.m.f.	magnetomotive force
m.p.	melting point
mV.	millivolt(s)
mμ	millimicron = 1×10^{-9} m. = 10 Å.
N	normal (solution)
N.T.P.	normal temperature and pressure
O.H.	open-hearth ; oil-hardened
O.Q.	oil-quenched
oz.	ounce(s)
p.d.	potential difference
*p*H	hydrogen-ion concentration
p.p.m.	parts per million
r.p.m.	revolutions per minute
sec.	second(s)
sp.gr.	specific gravity
sq.	square
S.W.G.	standard wire-gauge
T.	tempered
temp.	temperature
V.	volt(s)
VA.	volt-ampère(s)
W.	watt(s)
Wh.	watt-hour(s)
W.G.	water-gauge
W.Q.	water-quenched
wt.	weight
wt.-%	weight per cent
yd.	yard(s)
μ	micron(s) = 1×10^{-6} m.
μg	microgramme(s) = 1×10^{-6} g.
μμ	1 millionth micron = 1×10^{-12} m. = 0·01 Å.
Ω	ohm(s)
°	degree (arc or temperature)
′	minute of arc ; foot (feet)
″	second of arc ; inch(es)
<	less than
>	greater than
≮	not less than
≯	not greater than
≤	equal to or less than
≥	equal to or greater than
≠	not equal to
≡	identically equal to
≈	approximately equal to
∝	proportional to

Unless otherwise stated above, the symbols given in the following British Standards are used in the *Journal*: Chemical, Thermodynamical, and Physico-chemical quantities: B.S. 813; Engineering units: B.S. 560.

HUGH M. HENDERSON

HUGH M. HENDERSON was born in Glasgow in 1897. After receiving his education at Glasgow High School he entered the works of the Vauxhall and West Hydraulic Engineering Company, at Luton, and trained as a mechanical engineer from 1913 to 1917.

During the First World War Mr. Henderson served for a time in the Royal Air Force, and thereafter joined the Frodingham Iron and Steel Company. He has remained ever since with that Company, now renamed the Appleby-Frodingham Steel Company, a branch of The United Steel Companies, Limited, of Sheffield.

His work was at first concerned with maintenance and the erection of new plant, as a member of the Engineering Staff. In 1921 he transferred to the production side in the Rolling Mills and acted as Mill Manager from 1924 to 1928. In 1929 he again transferred his interests, this time to the commercial side of the Company's activities, and he took over control of the Frodingham Constructional Department in 1933. He was appointed Commercial Manager and a Director of the Appleby-Frodingham Steel Company in 1945.

Notwithstanding his later commercial activities Mr. Henderson has kept in close touch with technical developments in the industry. He has been a member of The Iron and Steel Institute from 1934. He has also been a member of the Lincolnshire Iron and Steel Institute from its early days and became its President in 1948.

Navana Vandyk

Hugh M. Henderson
Honorary Member of Council

IRON AND STEEL
INSTITUTE PAPERS

PLANT DESIGN · MANUFACTURE · RESEARCH · DEVELOPMENT

Carbide Distribution as a Possible Cause of Temper Brittleness

By L. D. Jaffe, Sc.D.

SYNOPSIS

The difficulties of reconciling current theories of temper brittleness of steel with the experimental data are pointed out. A new hypothesis is suggested : that temper brittleness may arise from the effect of temperature, in the sub-critical range, upon the quasi-equilibrium distribution of carbides or other phases. This hypothesis is examined theoretically, and it is found that the interfacial energy between ferrite and the phases mentioned would have to be approximately half the interfacial energy between grains of ferrite. Consequently, if the hypothesis is valid, the phase responsible for temper brittleness cannot be cementite.

Introduction

In examining possible causes of temper brittleness of steel, it is worth while to note that on the basis of published data[1] the embrittlement must arise from equilibrium or quasi-equilibrium considerations, and cannot depend on the kinetics of a reaction. Thus, the degree of embrittlement developed on long isothermal holding apparently depends only on the temperature of holding, and not on prior treatment. Specimens previously unembrittled and specimens previously severely embrittled approach the same intermediate degree of embrittlement when held at the same temperature.[1]

The explanation generally accepted for temper brittleness is that the phenomenon arises from precipitation from ferrite of iron nitride[2] or iron carbide,[3] precipitation taking place preferentially at grain boundaries. The nitrogen content of commercial steels is of the order of 0·005 to 0·010%. Such percentages of nitrogen appear to be soluble in iron (in equilibrium with iron nitride) above 300° C.[4–6] It is therefore somewhat difficult to understand how nitride precipitation could be associated with temper brittleness, which develops at 350–650° C. As this brittleness appears to develop in plain carbon steels as well as in alloy steels,[7] possible reduction of nitrogen solubility by alloying elements does not seem pertinent.

The carbon content of steels in which temper brittleness has been commonly observed is over 0·10% ; the solubility of carbon in ferrite (in equilibrium with cementite) is less than 0·02%, below 700° C.[4] A large quantity of excess iron carbide (cementite) is therefore always present during the development of temper brittleness. It is somewhat difficult to understand why any additional small quantity of carbide precipitating from the ferrite should not deposit on the pre-existing carbide particles, which provide nuclei for such a reaction, rather than at the grain boundaries. Precipitation on pre-existing particles would merely increase their size slightly and could hardly acount for the brittleness observed.

Another view of temper brittleness attributes it to equilibrium segregation of carbon, dissolved in the ferrite, in regions adjacent to grain boundaries.[8] On this basis it is rather difficult to explain the marked retardation of the development of temper brittleness which is caused by alloying elements.[7, 9] Alloying elements have little effect upon the rate of diffusion of carbon in austenite,[10, 11] and it seems likely that they have little effect upon the rate of diffusion of carbon in ferrite.

Because of these difficulties with existing theories it seemed worth while to consider a new hypothesis, which is, in a sense, a synthesis of the theories just mentioned : temper brittleness is due to *equilibrium segregation* of *carbide particles* at grain boundaries.

DISCUSSION OF HYPOTHESIS

It appeared possible that, besides the particles of carbide microscopically visible within the ferrite grains of tempered steel, there might be thinner particles concentrated at grain boundaries. These

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Dr. Jaffe is head of the Physical Metallurgy Section of Watertown Arsenal Laboratory, Watertown, Massachusetts, U.S.A.

particles might be in equilibrium with those within the grains, even though somewhat smaller, because their location at grain boundaries would reduce the associated interfacial energy. At low temperatures, because of the interfacial energy, the equilibrium distribution might be one in which the grain boundaries were covered with carbide, leading to brittleness. At higher temperatures, entropy would be more important, and a more random distribution of phases would be stable. In consequence, the equilibrium distribution might be one in which the concentration of carbides at grain boundaries was considerably less, leading to less brittleness. This would correspond to experimental data, for the degree of temper brittleness developed on long holding seems to be less the higher the temperature of holding.[1]

A basis for quantitative evaluation of this hypothesis was worked out in a recent report by the author on equilibrium distribution of a disperse phase between grain boundaries and grain interiors of a second, polycrystalline, phase.[12] It was found that temperature will significantly affect the distribution only if one of two conditions holds. The first condition is that γr^2 be of the order of kT, where, in the present application,

γ = interfacial energy per unit area between two ferrite grains
r^2 = volume of a carbide particle located at the ferrite grain boundary
k = Boltzmann's constant = $1{\cdot}380 \times 10^{-16}$ erg/° K.
T = absolute temperature.

The interfacial energy between ferrite grains has not been measured, nor has the energy between grains of any material. For several metals it appears[13] that " the grain-boundary energy is a very small fraction of the surface energy." Let it be assumed for the moment that the grain-boundary energy is of the order of $\frac{1}{1000}$ of the surface energy. Unfortunately the surface energy of solid iron also is not known, but it is probably of the same order as the surface energy of *liquid* iron,[14] 840 ergs/ sq. cm. On this basis, γ will be of the order of 1 erg/sq. cm. The amount of temper embrittlement developed on long holding is markedly affected by temperature changes when the temperature is near 600° C., so T may be taken as approximately 1000° K.

For $\gamma r^2 \sim kT$, we then must have $r^2 \sim 1{\cdot}4 \times 10^{-13}$ sq. cm. corresponding to a size for the grain-boundary particles of about 40 Ångstrom units. As the carbide particles within the grains of steels tempered above 600° C. appear to be resolvable under the optical microscope, their size is over 1000 Å. Since the assumed grain-boundary particles are much smaller than those within the grains, mininum energy would be attained by having zero concentration at the grain boundaries, and at low temperatures this condition would be approached. At high temperatures (above 600° C.), entropy would become important and the grain-boundary concentration would increase. This behaviour is the reverse of that originally postulated. The same conclusion will be reached for all values of γ down to 10^{-3} erg/sq. cm. which seems unreasonably low. The condition $\gamma r^2 \sim kT$ would therefore be of no help in accounting for temper brittleness.

The second condition under which the temperature may significantly affect the distribution of phases, is that the interfacial energy between ferrite and carbide should be approximately half the interfacial energy between two ferrite grains. When the ratio of these interfacial energies is one-half or less, most of the grain-boundary area will, under equilibrium conditions, be covered with particles of carbide in the form of very thin films.[12] These films (' network ') at the grain boundaries would be likely to make the material brittle.[15] If γr^2 is large compared with kT, as seems certain from the preceding discussion, then an increase in the ratio of interfacial energies to slightly over 0·5 would completely change the distribution of phases, lowering the concentration of the carbide at grain boundaries to practically zero, provided only that the grain-boundary particles are small in volume compared with the particles within the grains. For example, if the size (cube root of the volume) of grain-boundary particles is $\frac{1}{10}$ that of particles within the grains, a change in ratio of interfacial energies from 0·5 to 0·515 will reduce the grain-boundary concentration from over one-half to almost zero.[12] For a size ratio of $\frac{1}{100}$ the energy ratio need change only from 0·5 to 0·5005 ; for smaller grain-boundary particles, the energies need change even less. Certainly, an increase in temperature from 350° to 650° C. could increase the ratio of interfacial energies from 0·5 to 0·515, removing the grain-boundary network at equilibrium, and consequently producing a microstructure that would be much less brittle. It also seems certain that the grain-boundary particles, if there are any, must be small compared with the carbide particles within the grains. Thus, studies of the path of brittle fracture[8,16–18] in temper-brittle steels have established that the embrittlement is associated with grain boundaries, and metallographic etchants have been developed which produce a corresponding etching attack at the boundaries,[8,19–21] but grain-boundary particles have not yet been revealed by any metallographic technique.

A somewhat similar mechanism has been proposed[22] for embrittlement of copper by bismuth, though objections to the proposal have been pointed out.[15, 23, 24] As far as temper brittleness is concerned, the obvious objection to such a mechanism is that the ' carbide ' phase is unidentified. This could hardly be cementite, since the interfacial tension between cementite and ferrite is 0·93 of the interfacial tension between ferrite grains (at 690° C.).[15] The unknown phase might be some other carbide, or might not be a carbide at all but, instead, some other compound such as a phosphide.

CONCLUSIONS

Differences in quasi-equilibrium distribution of carbides or other phases at different sub-critical temperatures might give rise to temper brittleness.

A necessary condition would be that the interfacial energy between ferrite and these phases is approximately half the interfacial energy between grains of ferrite.

These carbides or other phases cannot be cementite. No suggestion as to their identity is made at this time.

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HISTORICAL NOTE No. 15

The Pinners and Wire-Drawers of York, 1296–1762

By Dr. H. R. Schubert

THE organized power of the crafts was " the most striking feature in the great development of civic life in which lay the main contribution of the Middle Ages to the cause of Western progress, and which reached its culmination about the middle of the fourteenth century."[1]

In this period, an Ordinance and Minute Book of the pinners and wire-drawers of York has its commencement ; its last entry is of 1762.[2] The initial sentence of the book formally proclaims that it was made in 1349, but was corrected and augmented in the second period of the mayoralty of Thomas Harrison who, on 6th October, 1592, " diligentlye perused, reformed and enlargened " the " anciente ordinannce " (*i.e.* charter) of the craft. This charter, in its form of 1592, occupies the major part of the Ordinance and Minute Book (folios 3 to 24). It is a codification of all the rules binding for the pinners and wire-drawers of York at the date of its enactment. In the first part, it also reinforces many rules made since 1349.

Originally, the pinners and wire-drawers at York were two separate crafts. A wire-drawer is first referred to in the Register of the Freemen of the City of York, in 1300.[3] Evidently the pinners emerged from the ' nedelers ' or ' aiguilers ' mentioned in the same Register from 1296 to 1339, after which date these terms completely disappeared from the Register, and were replaced by the term ' pinner ' from 1349 onwards, *i.e.*, the very year in which the Ordinance and Minute Book begins.[4] Pinners and wire-drawers were not combined into one craft guild before the second period of the mayoralty of Richard Yorke, *i.e.*, in 1480–1481.[5] Consequently, all enactments inserted in the charter of 1592 and dated prior to 1480–81 apply to the pinners only. This is corroborated by an order of 6th September, 1421(inserted as paragraph 6) naming seventeen masters, thirteen of whom can be verified as pinners by entries in the Register of the Freemen, and by a petition of 31st January, 1422.[6]

Four paragraphs of the charter of 1592 preceding the order of 1421 mentioned already, evidently

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[1] G. Unwin, " The Gilds and Companies of London," p. 61, third edition. London, 1938.

[2] Guildhall at York, MS. No. 60. The author is greatly indebted to the City Archivist, the Rev. Angelo Raine, for drawing his attention to the manuscript.

[3] " Publications of the Surtees Society," vol. 96, p. 8. Durham and London, 1897.

[4] Publications, *loc. cit.*, vol. 96, pp. 6 and 44 : the terms " le Nedeler " and " le Aguiller " were used at Chichester in the thirteenth century, Victoria County History of Sussex, vol. 3., p. 97. London 1935.

[5] Publications, *loc. cit.*, vol. 125, p. 297. Durham and London, 1915.

[6] " Publications, *loc. cit.*, vol. 96, *passim*, vol. 125, p. 103.

constitute remnants of a charter of 1349, which is now lost and can only be reconstructed from these four paragraphs. They are concerned with rules for apprenticeship and quality of ware. The articles manufactured had to conform to the length of certain samples given out by the 'searchers' of the craft, who controlled the accuracy of work. No master was allowed to employ any stranger unless he was apprenticed to the artifice and his name was entered in the ordinance book of the craft. The period of apprenticeship was fixed at eight years, which was reduced in a later, but undated, ordinance to seven years (the usual period in the Middle Ages) and, subsequently, to six years.

The sixth paragraph contains the order of 6th September, 1421, compelling all members of the craft to submit to the control of the 'searchers.' The next part reinforces an older enactment, dated 23rd October, 1425, protecting the craft against competition from outside, particularly from " any alyen of any other nacion," by an extraordinarily high fine of 40*s*. imposed in each case of contravention. It is followed by an injunction of 14th December, 1582, ordering that foreign pinners coming to York to sell their pins should resort to the Thursday markets, and refrain from any attempt to sell by hawking.

The remaining part of the charter of 1592 (paragraphs 12 to 28) consists of more prescriptions regulating trade and activities of the craft. In particular, strict control by the 'searchers' was again imposed, especially in respect of ware offered for sale by foreigners. Further, no master was allowed to take on a second apprentice before the first apprentice had served him for at least five years. It also was strictly forbidden to work on Sundays and festival days, and to visit " any tavern, inn, or ale house " during the time of " divyne service or sermons." To ensure work of good quality, every member of the craft wanting to set up as a master had to be examined by the 'searchers' as to his ability, and if he was unable to pass the test, he had to continue working as a journeyman.

Of particular interest, is the variety of finished goods manufactured by the pin-makers and wire-drawers of York. Pins, the small iron teeth inserted in the cards used for wool-carding, and fish-hooks were the main products these York craftsmen had in common with similar craftsmen in Bristol, Coventry, and London. In an ordinance of 1480–81 shoe buckles were added to the articles manufactured by the craft.[7] In the enactment of 1582 mentioned above, reference is made to bread grates, mice traps, and " heckles of iron wire." *i.e.*, hackling-combs used for flax-dressing, the teeth of which were tinned over and set in dry wood without sap.[8]

In 1607, it was ordained it should be lawful for any member of the craft to make malt-creels " as well the wood worke thereof as the wyer worke of the same as heretofore."[9] In addition, in an ordinance of 1st September, 1619, the following articles manufactured and sold by the pinners and wire-drawers of York are enumerated : bird cages " being part made of wyer," wire trellises, swine rings, knitting needles, network needles, pack needles, key chains, surveying chains, buttons of brass, steel, or tin, clasps, keepers pins, dripping pans made of plate, lanterns made of plate, spice grates, meat broylers, block candlesticks, tinder boxes, plate measures of all sorts, funnels and watering pots of plate, or " any other plate worke."[10]

At York, the pinners always had predominance in numbers over the wire-drawers. This is indicated by the Register of Freemen of the City of York, with 28 names of pinners from 1349 to 1398, but 16 names of wire-drawers in the same period. In the first half of the following century this is still more significant, with 32 pinners, but four wire-drawers and only one wire-smith. After the combination of the two crafts the proportion was not less unfavourable to the wire-drawers. For example, in 1593, the wire-drawers appear to be completely outnumbered by the pinners, nine of the ten masters and 'searchers' mentioned being pinners.[11] The predominance of the pinners was probably caused by the demand for pins and needles by the numerous tailors of York, whose guild was very much larger than that of any of the City's metal-working guilds, numbering one hundred and twenty-eight master tailors in 1386, whereas none of the guilds of metal workers reached a membership of twenty in the same period.[12]

The comparatively small number of wire-drawers at York is probably due to two main causes. In the fifteenth and sixteenth centuries, York had extensive trade with the Low Countries, where wire was manufactured and exported in large quantities.[13] In addition, competition from Coventry, the principal centre of medieval wire-drawing and wool-cardmaking in England, was strongly felt. Early in the fifteenth century, importation of " cardleves " from Coventry into York was prohibited.[14] There are are also signs of infiltration from Coventry into the York wire trade ; in 1363 and 1368, two wire-drawers at York are designated in the Register of the Freemen[15] as being from Coventry (*de Coventre*).

An interesting item is the purchase of wire between 1484 and 1544 for the Fabric of York Minster.[16] The names of two York wire-drawers supplying such wire (at 12*d*. a pound) are mentioned in the Fabric Rolls: Ricard Symson in 1484–85, and Johannes Jarrot in 1525–26. The wire supplied was used for mending the clockwork.

[7] " Publications of the Surtees Society, vol. 125, p. 297. Durham and London, 1915.
[8] Guildhall at York, MS. No. 60, fol. 8.
[9] *Ibid.*, fol. 16^{v}–17.
[10] *Ibid.*, fol. 17^{v}.
[11] *Ibid.*, fol. 16.
[12] " Victoria County History of Yorkshire," vol. 2, p. 391. London, 1912.
[13] Publications, *loc cit.*, vol. 129, pp. XXII, 71, 72, 104. Durham and London, 1918.
[14] The prohibition order is published in Publications, *loc cit.*, vol. 120, p. 79 (Durham and London, 1912), but without a date. The names of the York cardmakers mentioned in the order compared with the names in the Register of Freemen, suggest a date of about 1420.
[15] Publications, *loc cit.*, vol. 96, pp. 59 and 67.
[16] Minster Library at York, Fabric Rolls. The author is greatly indebted to the Librarian, Canon Harrison, for facilitating the study of these valuable account rolls.

Tessellated Stresses – Part V

By F. László, Dr. Ing.

SYNOPSIS

The currently accepted theory of the influence of strains due to volume change on the shape of precipitates is carefully examined but rejected. The rules giving the true shaping effort of these strains, in terms of the elastic constants, and the yield stress of matrix and precipitate, are determined.

Experimental observations with ferrous and non-ferrous metals are reviewed in the light of the new theory.

The conventional views on dislocation are extended by the idea that the atomic structure must become buckled. This leads to suggestions as regards the interpretation of hardness phenomena.

Some special questions discussed in previous parts are reconsidered and explained in greater detail.

Introductory Remarks

For the sake of continuity with the author's previous publications on tessellated stresses, *viz.*, Parts I to IV, the numbers of equations, tables, and diagrams given in this paper are continued from Part IV. References to equations, tables, diagrams, etc., which appear as follows in Parts I to IV are given the appropriate suffix [I], [II], [III], or [IV]:

	Equations	Tables	Diagrams
Part I*	(1) to (7)	I to III	Figs. 1 to 7
Part II†	(8) to (21)	IV to IX	Figs. 8 to 15
Part III‡	(22) to (36)	X to XV	Figs. 16 and 17
Part IV§	(37) to (67)	XVI to XIX	Figs. 18 to 20

List of Symbols

The notation of the present paper is mostly in accordance with that used in Parts I to IV. Certain modifications which appear to be desirable, and most of the new symbols, are given in the following list:

(U) = total strain energy associated with a precipitate of unit volume embedded in an infinite matrix.

$(U)_w$ = strain energy with unyielding cylindrical precipitate in unyielding matrix.

$(U)_{uyi}$ = strain energy with yielding lamellar precipitate in unyielding matrix.

$(U)_{vyo}$ = strain energy with unyielding spherical precipitate in yielding matrix.

$(U)_{wyio}$ = strain energy with yielding cylindrical precipitate in yielding matrix.

β_{uyi}, β_{wyo} = strain energy ratios expressing the quotient of $(U)_{uyi}/(U)_u$ and $(U)_{wyo}/(U)_u$ respectively.

β_{wyio} = strain energy ratio expressing the quotient of $(U)_{wyio}/(U)_u$.

β'_{wyi}, β'_{vyo} = strain energy ratios expressing, for precipitate and matrix having the same elastic constants, the quotients of $(U)_{wyi}/(U)_u$ and $(U)_{vyo}/(U)_u$ respectively.

f_{yo} = yield stress of matrix.

f_{yi} = yield stress of precipitate.

α_i, α_o = yield-stress parameters as defined by:

$$f_{yi} = \alpha_i \frac{m_i E_i}{m_i - 1}\lambda \text{ and } f_{yo} = \alpha_o \frac{m_o E_o}{m_o - 1}\lambda$$

VOLUME CHANGE DURING NASCENT PRECIPITATION

F. R. N. Nabarro[1, 2] attempted the analysis of the factors controlling the shapes of precipitates and the hardening effect. As regards the shape of nascent precipitates, he calculated firstly the strain energy of a monocrystalline spherical precipitate[1] to be for unit volume of precipitate of a cubic metal:

$$(U)_s = \frac{3}{2}(c_{11} + 2c_{12})\lambda^2 = \frac{3}{2}\frac{\lambda^2}{s_{11} + 2s_{12}} \quad \ldots\ldots\ldots(68a)$$

where the constants c_{11}, c_{12}, s_{11}, s_{12} correspond to those of equations (24) and (25) [III], and λ is the linear volume change by fraction, *i.e.*, the strain potential is $-\lambda$ in accordance with p. 182P of Part I, or in Nabarro's wording "the (linear) misfit." The above crystal constants refer of course to the material of the precipitate.

Consider equation (30) [III], which is valid for any cubic material. Thus equation (68*a*) for the monocrystalline spherical precipitate may also be expressed in terms of the elastic constants for the random aggregate of the precipitate:

$$(U)_s = \frac{3}{2}\frac{m_i E_i}{m_i - 2}\lambda^2 \quad \ldots\ldots\ldots\ldots\ldots(68b)$$

Consider also a unit sphere of the random aggregate of the crystals of the precipitate, having external radius r_j and surface area $3/r_j$. If it is intended to suppress the would-be change in volume by 3λ, a radial hydrostatic stress of:

$$p_j = -\frac{m_i E_i}{m_i - 2}\lambda$$

is to be applied to the sphere. The stress inside the precipitate will be, at any point and in every direction, a pure normal stress of the above value of p_j. Hence the strain energy per unit volume of a random spherical precipitate will be:

$$\frac{1}{2}p_j\frac{3}{r_j}(-r_j\lambda) = \frac{3}{2}\frac{m_i E_i}{m_i - 2}\lambda^2;$$

that is, exactly the same as for the monocrystalline precipitate.

It should be noted that the monocrystalline spherical precipitate has been supposed[1] to have a lattice exactly registering with that of the matrix. The above spherical random precipitate cannot in general register with the matrix. However, if it is imagined that a monocrystalline spherical precipitate is rotated in its cavity and thus the lattices of matrix and

* *Journal of The Iron and Steel Institute*, 1943, No. I, p. 173P.

† *Ibid.*, 1943, No. II, p. 137P.

‡ *Ibid.*, 1944, No. II, p. 183P.

§ *Ibid.*, 1945, No. II, p. 207P.

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precipitate disregister, while the cohesion between the two is maintained over misfit atoms, there is no reason whatsoever to expect any change in the strain energy of the rotated precipitate beyond quantities of minor order of magnitude. Thus let it be concluded that the strain energy of the above-discussed nature of a monocrystalline spherical precipitate is the same whether or not the lattices of precipitate and matrix register, and is identical with that of a hypothetical spherical random precipitate.

A calculation[1] has been made of the strain energy of a monocrystalline lamellar precipitate of a cubic metal, the lattice of which registers with that of the (cubic) matrix, the plane of the lamella being a (100) crystal interface. The strain energy per unit volume of precipitate for a volume change of the precipitate by 3λ amounts to[1] :

$$(U)_l = \frac{(c_{11} + 2c_{12})(c_{11} - c_{12})}{c_{11}}\lambda^2 = \frac{\lambda^2}{s_{11} + s_{12}} \quad \ldots\ldots(69)$$

corresponding to equation (66) [IV]. The elastic crystal constants refer, of course, to the precipitate. Equation (69) is based on the consideration that the change in thickness by λ does not involve the development of stresses across the lamella and that the would-be change in length in the plane of the precipitate is completely suppressed solely by the elastic strains of the precipitate in its plane.

The above equation (69) may also be obtained by considering a 'free' monocrystalline right rectangular lamella of precipitate of unit volume, with edges parallel to the crystal axes, to be subjected to 'external stresses' $f_x \neq 0$ and $f_y \neq 0$, in the plane of the lamella, and $f_z = 0$ across it, so that the edges parallel to f_x and f_y change in length by $-\lambda$. Then, with equation (25) [III] :

$$-\lambda = s_{11}f_x + s_{12}f_y = s_{12}f_x + s_{11}f_y\,;\, f_x = f_y = f_{xy} = -\frac{\lambda}{s_{11} + s_{12}}$$

and the strain energy per unit volume is :

$$2\tfrac{1}{2}(-\lambda)f_{xy} = \frac{\lambda^2}{s_{11} + s_{12}}.$$

This consideration will be of assistance in some later decision.

If the effect of disregistry is to be investigated on a lamella which is cohesively embedded in the matrix, equation (8) [II] makes a precise analysis possible. The disregistry will be amply represented if the unit cubes of Fig. 8 [II] are built up of monocrystalline lamellae of a cubic precipitate metal, with crystal axes parallel to the edges of the unit cubes, and of cohesively jointed alternative lamellae of random aggregates of the matrix crystals. The random aggregate of unit cubes is, of course, also cohesively jointed. For the matrix lamellae the elastic constants for random aggregates, *viz.*, E_o and $1/m_o$, are to be considered, while for the precipitate, $1/E$ is to be replaced by s_{11}, and $1/mE$ by $-s_{12}$, f_2 by f_{xy}, and for the whole cube f_q by f_z. In this case only the precipitate changes volume by the fraction 3λ, which, however, causes in general also a change in volume of the whole lamellar unit cube by some fraction $3\lambda'$. Then the previous equation (8) is to be modified as follows :

$$\left.\begin{aligned} \lambda' &= \frac{m_o - 1}{m_o E_o} f_o - \frac{f_z}{m_o E_o} \\ \lambda' &= \lambda + (s_{11} + s_{12})f_{xy} + s_{12}f_z \\ \lambda' &= u_i\lambda + \left(u_i s_{11} + \frac{u_o}{E_o}\right)f_z + 2u_i s_{12}f_{xy} - \frac{2u_o f_o}{m_o E_o} \\ 2(u_i f_{xy} + u_o f_o) &+ f_z = 0 \end{aligned}\right\} \quad \ldots(70)$$

The above equations represent an approximation only, in general, but lead to exact results if, for example, the volume fraction of one of the components approaches nought, as was pointed out on p. 184P [III] and on p. 221P [IV]. In the case of nascent precipitation, $u_i = 0$ is to be considered, so that :

$$f_z = 0, f_o = 0, \lambda' = 0, \text{ and } f_{xy} = -\frac{\lambda}{s_{11} + s_{12}}.$$

By applying the method outlined on p. 221P [IV] and using equation (19) [II], the strain energy per unit volume of such nascent monocrystalline lamellar precipitates is obtained as :

$$\left[\frac{dU}{du_i}\right]_{u_i = 0} = \frac{\lambda^2}{s_{11} + s_{12}} \quad \ldots\ldots\ldots\ldots(71)$$

The strain energy expressed by equation (71) cannot be affected if the matrix lamellae are considered to be monocrystalline, the crystal axes of which are not parallel with the edges of the unit cube in order to provide for disregistry. Though equation (70) would require very complicated modification, the strain energy would be again expressed by equation (71) because it is obviously independent of the elastic properties of the matrix. Since equations (69) and (71) are identical, it is to be concluded that the strain energy of a monocrystalline lamellar precipitate of the above definition is independent of whether the lattices of matrix and precipitate register or not. This is in contradiction to certain views.[1, 2]

Let equations (64*a*) to (64*c*) [IV] be remembered. If precipitates and matrix were quasi-isotropic, these equations would express the strain energy of nascent precipitates when " reduced to unit volume," owing to the change in volume of the precipitate by 3λ. Let the following notation be introduced for the strain energy per unit volume of lamellar, cylindrical, and spherical quasi-isotropic precipitates, respectively :

$$(U)_u = \left[\frac{dU}{du_i}\right]_{u_i = 0} = \frac{m_i E_i}{m_i - 1}\lambda^2 \quad \ldots\ldots\ldots\ldots(72a)$$

$$(U)_w = \left[\frac{dU}{dw_i}\right]_{w_i = 0} = \frac{\frac{3}{2}\frac{m_i + 1}{m_i E_i} + \frac{1}{2}\frac{m_o + 1}{m_o E_o}}{\frac{m_i + 1}{m_i E_i}\frac{m_i - 2}{m_i E_i} + \frac{1}{E_i}\frac{m_o + 1}{m_o E_o}}\lambda^2 \quad (72b)$$

$$(U)_v = \left[\frac{dU}{dv_i}\right]_{v_i = 0} = \frac{3}{2\frac{m_i - 2}{m_i E_i} + \frac{m_o + 1}{m_o E_o}}\lambda^2 \quad \ldots\ldots(72c)$$

The equations (72*a*) to (72*c*) account for the sum of strain energies developing in precipitate and matrix. Table XX shows, in addition to the total strain energy, the individual contributions by matrix and precipitate, in rows 1 to 3 for the general case, and in rows 4 to 6 for incompressible precipitate, *i.e.*, $m_i = 2$.

It is noticed that the strain energy associated with lamellar random precipitates develops in the precipitate only, without the matrix taking any share in it, reciprocating the result obtained for monocrystalline

Table XX

STRAIN ENERGY ASSOCIATED WITH PRECIPITATES OF UNIT VOLUME HAVING LAMELLAR, CYLINDRICAL, OR SPHERICAL SHAPE

Row	Strain Energy	Shape of Precipitate		
		Lamellar	Cylindrical	Spherical
1	Total	$\frac{m_i E_i}{m_i - 1}\lambda^2$	$\frac{\frac{3}{2}\frac{m_i+1}{m_i E_i} + \frac{1}{2}\frac{m_o+1}{m_o E_o}}{\frac{m_i+1}{m_i E_i}\frac{m_i-2}{m_i E_i} + \frac{1}{E_i}\frac{m_o+1}{m_o E_o}}\lambda^2$	$\frac{3}{2\frac{m_i-2}{m_i E_i} + \frac{m_o+1}{m_o E_o}}\lambda^2$
2	Precipitate	$\frac{m_i E_i}{m_i - 1}\lambda^2$	$\frac{\frac{1}{2}\frac{m_i-2}{m_i E_i}\left[3\left(\frac{m_i+1}{m_i E_i}\right)^2 + 2\frac{m_i+1}{m_i E_i}\frac{m_o+1}{m_o E_o}\right] + \frac{1}{2E_i}\left(\frac{m_o+1}{m_o E_o}\right)^2}{\left(\frac{m_i+1}{m_i E_i}\frac{m_i-2}{m_i E_i} + \frac{1}{E_i}\frac{m_o+1}{m_o E_o}\right)^2}\lambda^2$	$\frac{6\frac{m_i-2}{m_i E_i}}{\left(2\frac{m_i-2}{m_i E_i} + \frac{m_o+1}{m_o E_o}\right)^2}\lambda^2$
3	Matrix	Nought	$\frac{\left(\frac{m_i+1}{m_i E_i}\right)^2\frac{m_o+1}{m_o E_o}}{\left(\frac{m_i+1}{m_i E_i}\frac{m_i-2}{m_i E_i} + \frac{1}{E_i}\frac{m_o+1}{m_o E_o}\right)^2}\lambda^2$	$\frac{3\frac{m_o+1}{m_o E_o}}{\left(2\frac{m_i-2}{m_i E_i} + \frac{m_o+1}{m_o E_o}\right)^2}\lambda^2$
4	Total for $m_i=2$	$2E_i\lambda^2$	$\left(\frac{E_i}{2} + \frac{9}{4}\frac{m_o E_o}{m_o+1}\right)\lambda^2$	$3\frac{m_o E_o}{m_o+1}\lambda^2$
5	Precipitate for $m_i=2$	$2E_i\lambda^2$	$\frac{E_i}{2}\lambda^2$	Nought
6	Matrix for $m_i=2$	Nought	$\frac{9}{4}\frac{m_o E_o}{m_o+1}\lambda^2$	$3\frac{m_o E_o}{m_o+1}\lambda^2$

lamellar precipitates. Let a ' free ' lamellar random precipitate change its volume by the fraction 3λ, and then be forced back to the unstrained size, except for its thickness, across which there should be considered to be no stress. The requisite principal stresses in the plane of the lamella are both :

$$-\frac{m_i E_i}{m_i - 1}\lambda$$

and the strain energy which develops per unit volume is found to be :

$$2\tfrac{1}{2}(-\lambda)\left(-\frac{m_i E_i}{m_i - 1}\lambda\right) = \frac{m_i E_i}{m_i - 1}\lambda^2;$$

that is, exactly the same as given by equation (72*a*). Thus it is noticed that, as regards stresses and strain energy, monocrystalline and quasi-isotropic lamellar precipitates have characteristic features which are identical.

As regards spherical random precipitates in random matrix, row 3 of Table XX indicates that the strain energy of the matrix would not become nought unless :

$$\frac{m_o + 1}{m_o E_o} = 0.$$

The latter condition could be satisfied only by $E_o = \infty$, which is tantamount to absurdity. It must then be concluded that there is no justification either for assuming that the strain energy developing in the matrix around a monocrystalline spherical precipitate would ever become negligibly small. Thus the values suggested by equations (68*a*) and (68*b*), for the total strain energy associated with monocrystalline spherical precipitates, must be wrong, since no share is taken by the matrix.

Equations (68) and (69) have been compared and the conclusion has been drawn[1] that, in general, the strain energy of a nascent lamellar precipitate is much less than that of a spherical one. Since equation (68) must be discarded when the question is to be answered as to which shape of nascent precipitate would be accompanied by the least strain energy due to the volume change of the precipitate, the following procedure is suggested.

As the correct method of calculation of the strain energy of minute monocrystalline spherical precipitates embedded in a monocrystalline matrix is at present lacking, an approximation should be made by considering random aggregates of different shapes of precipitate embedded in random aggregates of matrix. This method of approximation may be examined to some extent, since precise values are at hand for the strain energy of lamellar precipitates of both monocrystalline and quasi-isotropic structure, *viz.*, equations (69) and (72*a*) respectively.

Table XXI shows in columns 1 and 2 the values of $(U)_l$ and $(U)_u$ for all the cubic metals considered in Table XI [III], calculated for $\lambda = 0{\cdot}001$. The ratio of $(U)_l$ and $(U)_u$ is shown in column 3. With elastically isotropic crystals $s_{44} = 2(s_{11} - s_{12})$ ought to be

Table XXI

STRAIN ENERGY OF MONOCRYSTALLINE AND QUASI-ISOTROPIC LAMELLAR NASCENT PRECIPITATES FOR $\lambda = 0{\cdot}001$, AND THE DEGREE OF ELASTIC ISOTROPY OF THEIR CRYSTALS

	$(U)_l$, in. lb./cu. in.	$(U)_u$, in. lb./cu. in.	$(U)_l/(U)_u$	$\frac{s_{44}}{2(s_{11} - s_{12})}$
	(1)	(2)	(3)	(4)
Iron	30·53	42·94	0·711	0·415
Sodium	0·6306	1·912	0·330	0·133
Potassium	0·3202	0·7929	0·404	0·159
Aluminium	14·36	15·43	0·931	0·810
Copper	16·67	26·52	0·629	0·312
Brass (72/28)	13·13	22·83	0·575	0·251
Silver	10·93	17·54	0·623	0·346
Gold	10·99	19·60	0·561	0·349

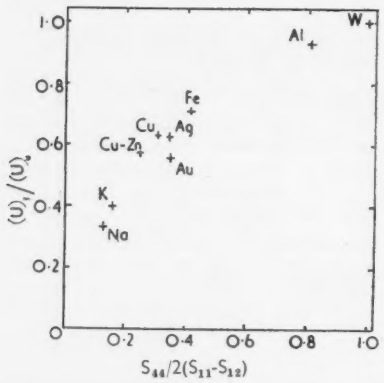

Fig. 21—Correlation between the degree of elastic isotropy, and the ratio of the strain energy of monocrystalline and random lamellar precipitates

observed. Thus the actual degree of isotropy of a cubic crystal may be indicated by $s_{44}/2(s_{11} - s_{12})$, the value of which, calculated with the figures of Table XI [III], is tabulated in column 4 of Table XXI. In Fig. 21 the values of column 3 are plotted against those of column 4, and the points are marked with the symbol of the elements concerned, the notation Cu–Zn referring to 72/28 brass. For tungsten, the crystals of which are perfectly isotropic, both the abscissa and the ordinate are equal to 1·0. This point also is plotted. It is seen that the ratio $(U)_l/(U)_u$ appears to be controlled by the degree of elastic isotropy. There is no reason to assume that conditions might be different in the case of cylindrical or spherical precipitates. The consideration of hypothetical precipitates of random crystal structure in hypothetical matrices, also of random crystal structure, is accompanied by the additional advantage that the combination of metals of any type of crystal may be investigated.

Under these circumstances, equation (65) [IV] applies to nascent precipitation, stating that by the strain energy involved, the deposition of lamellar precipitates should be fostered if the shear modulus of elasticity of the precipitate G_i is less than that of the matrix G_o, and that of spherical precipitates if $G_i > G_o$. Nascent cylindrical precipitates should not develop at all, if strain energy due to volume change of the precipitate is the controlling factor.

However, besides the strain energy involved, surface tension is actually a powerful shaping agent, invariably attempting to compact the precipitate. A third factor in the shaping of nascent precipitates is the mechanics of deposition. The process of precipitation apparently begins with atoms being rejected in certain minimum units of specified shape across certain crystal interfaces of the matrix. Also, the rate of rejection is a feature of the mechanics of deposition.

When using equations (68) and (69) for silver precipitate, $(U)_s = 65{\cdot}14 \times 10^6\lambda^2$ in.lb. and $(U)_l = 10{\cdot}93 \times 10^6\lambda^2$ in.lb. respectively. An attempt has been made[1] to correct equation (68). While for silver precipitates in a copper matrix, $(U)_s/(U)_l \approx 6$, it has been suggested[1] that this value should be corrected to about 3, since about half of the strain energy $(U)_s$ might be taken by the matrix. However, if $(U)_s$ with equation (68*a*) were the correct value of the strain energy associated with a hypothetical spherical nascent precipitate of silver embedded in a copper matrix, the whole of $(U)_s$ ought to be taken into consideration and compared with the whole of $(U)_l$. While $(U)_l$ is no doubt a correct value for a monocrystalline lamellar nascent precipitate of silver in a copper matrix, as regards $(U)_s$, a comparison suggests itself with $(U)_v = 23{\cdot}79 \times 10^6\lambda^2$ in.lb., of which the silver precipitate takes $8{\cdot}71 \times 10^6\lambda^2$ in.lb. and the copper matrix $15{\cdot}08 \times 10^6\lambda^2$ in.lb., all the latter values being calculated by the equations tabulated in Table XX. A lamellar random precipitate of silver would have, with Table XXI, $(U)_u = 17{\cdot}54 \times 10^6\lambda^2$ in.lb., all of which would be taken by the precipitate. Since $(U)_l < (U)_u$, lamellar precipitate is fostered, though the strain energy bound solely to the spherical precipitate, *viz.*, $8{\cdot}71 \times 10^6\lambda^2$ in.lb., is not much more than half of $(U)_u$. The ratio $(U)_v/(U)_u$ amounts, of course, only to 1·36 and not to something like 6, as has been suggested[1] for monocrystalline precipitates. The deviation should, however, be attributed predominantly to the procedure, not of distributing the strain potential $-\lambda$ between matrix and precipitate, but of imposing the whole of it exclusively upon the precipitate. If the whole strain potential $-\lambda$ is imposed upon a spherical random silver precipitate in a random matrix, *i.e.*, if $E_o = \infty$ is considered, instead of the correct value for the strain energy of $23{\cdot}79 \times 10^6\lambda^2$, a value of $65{\cdot}14 \times 10^6\lambda^2$ issues, in accordance with Table XX, corresponding to $(U)_s$ for the monocrystalline precipitate. Of course, the half of 65·14 is still much greater than 17·54, *viz.*, the factor of $(U)_u$. Thus the fallacious argument[1] could also be produced for quasi-isotropic matrix and precipitate.

VOLUME CHANGE DURING ADVANCING PRECIPITATION

The considerations of the preceding section were, without mentioning it, based on the consideration that the misfits involved by the volume change are balanced exclusively by elastic strains of precipitate and/or matrix. If this condition is maintained also during advancing precipitation, the effect of the strain energy remains the same as already mentioned. While equations (72*a*) to (72*c*) lose their validity for considerable volume fractions of precipitate, equation (63) [IV] is not limited in this regard. The latter states that if the shear modulus of elasticity of the precipitate is less than that of the matrix, gradually rejected atoms of the precipitate are urged to attach themselves to the actual precipitate so that its shape should become as little compact as possible; but if the conditions regarding shear modulus are opposite, compacting of the precipitate should be fostered. Of course, with nascent as well as with advanced precipitation, the shape of precipitate is not confined to the lamellar or spherical one. If conditions are considered, when, *e.g.*, $G_i < G_o$, and other effects are negligible, advancing growth of a precipitate must modify the nascent lamella to some oblate spheroid of some

degree of compactness, which is a further feature of the mechanics of deposition.

There have been different experimental observations which have stimulated the consideration that the primary process of precipitation becomes modified after some progress. It has been suggested[1, 2] that after some thickening of nascent lamellae, their lattice breaks away from that of the matrix in a simultaneous and concentrated process. The latter idea followed a suggestion of W. L. Bragg stating that, in the present author's wording, a disc-shaped cavity would accommodate, with negligibly small strain energy in the matrix, the change in volume of a 'lamellar' precipitate. Thus a calculation[2] was made of the strain energy developing individually in a quasi-isotropic matrix which contains cylindrical or spherical precipitates, the precipitate being assumed to be incompressible, *viz.*, $m_i = 2$. The results[2] are in perfect accordance with row 6 of Table XX. In addition to the previous argument[1] that any nascent precipitate must be lamellar since $(U)_l < (U)_s$, it was suggested that the lattice of the nascent precipitate must break away from that of the matrix as soon as the thickness of the precipitate had increased sufficiently for its strain energy with equation (69) to balance the energy required for the lattice separation, *viz.*, the heat of melting of a monatomic layer of the flat interfaces of the precipitate with the matrix. This suggestion was based on the argument that no strain energy is required for the accommodation of a lamellar precipitate if its lattice has disregistered. It has been exactly proved above that this latter argument is, in general, entirely wrong. On considering Table XX and the unquestionable principle that the joint strain energy of matrix and precipitate must be taken into account, it is recognized that the breaking away of the lattice could reduce the joint strain energy to nought, only if the precipitation is accompanied by a reduction in volume of the precipitate, and the cohesion between precipitate and matrix is completely destroyed by the 'breaking away.' If either of these conditions is not satisfied, the breaking away, in the above sense, could hardly ever reduce the strain energy, but it might slightly increase it in the case of the cohesion still being maintained, by creating misfit atoms. Consequently, the associated Fig. 1 of reference 1 and Fig. 2 of reference 2, being valid only for the individual strain energy of the matrix, are of no import in connection with the problem under consideration.

As a further question the influence of the strain energy due to volume change, on the shaping of broken-away precipitates, has been discussed.[2] This discussion was based on the analysis of the strain energy developing in an infinite matrix which surrounds a precipitate of the shape of a spheroid. The matrix was once more considered to balance the whole change in volume of the precipitate. It is seen, however, in rows 4 to 6 of Table XX that, also with an incompressible precipitate, the individual contribution of the precipitate to the strain energy total of the compound body is nought for a spherical precipitate, but it is not at all negligible with a cylindrical one, and that it accounts for the whole of the strain energy of the system with a thin lamella. Here, also, of course, the strain energy total, and not the individual contribution of the matrix, is the critical factor. The above-mentioned analysis was applied first of all to thin lamellae just when they broke away from the matrix, *i.e.*, to a case for which the strain energy of the matrix is nought, but that of the precipitate (which was wrongly assumed[2] also to be nought) amounts to $2E_i\lambda^2$ for an incompressible precipitate, according to Table XX. Hence, together with all considerations related to it, Fig. 3 of reference 2 is entirely wrong.

It is therefore suggested that future investigations of the type just discussed should consider, for spheroids of unit volume having polar diameter c and equatorial diameter a, that the strain energy for $c/a \leq 1$ is :

$$(U)_{uv} = \left[1 - \phi\left(\frac{c}{a}\right)\right](U)_u + \phi\left(\frac{c}{a}\right)(U)_v \quad \ldots\ldots(73a)$$

For $1 \leq c/a \leq \infty$ the strain energy is :

$$(U)_{wv} = \left[1 - \psi\left(\frac{a}{c}\right)\right](U)_w + \psi\left(\frac{a}{c}\right)(U)_v \quad \ldots\ldots(73b)$$

where both functions are required to satisfy the conditions of $\phi(0) = \psi(0) = 0$ and $\phi(1) = \psi(1) = 1$.

TESSELLATED YIELD DURING PRECIPITATION

It is suggested that tessellated yield is one of the clues to the question as to why an initial increase in hardness, owing to precipitation, might be modified in the course of advancing precipitation, and that it may explain other phenomena connected with precipitation, *e.g.*, certain changes in shape of precipitates. First, let the strain energy conditions be investigated for precipitates of the above three simple shapes, as to the possible reduction by yield. In order to obtain non-dimensional results, the yield stress of precipitate and matrix will be expressed respectively as :

$$f_{yi} = -\alpha_i \frac{m_i E_i}{m_i - 1}\lambda \text{ and } f_{yo} = \alpha_o \frac{m_o E_o}{m_o - 1}\lambda$$

where α_i and α_o are the yield-stress parameters. The different signs are necessary, since, for example, if the precipitate expands in volume, *viz.*, $3\lambda > 0$, 'compressive' yield should occur in the precipitate, if any, and 'tensile' yield in the matrix. The following cases will be analysed : (1) When the precipitate can yield and the matrix not at all, (2) for conditions opposite to those in (1), and (3) when both possess limited yield resistance only. It is most convenient to carry out the analysis for precipitates of unit volume embedded in an infinite matrix, and not for minute precipitates in unit matrix volume, the results being the same with either calculation.

Lamellar Precipitate of Limited Yield Resistance in Elastic Matrix

If in a given case the yield-stress parameter is less than unity, two principal stresses in the plane of a very thin lamella are equal to :

$$f_{yi} = -\alpha_i \frac{m_i E_i}{m_i - 1}\lambda$$

and the third principal stress across the lamella is negligible. The plastic strain in the plane of the lamella in every direction is :

$$s_{yi} = -\lambda - \frac{m_i - 1}{m_i E_i} f_{yi}.$$

Strain energy develops in the precipitate only, and it amounts per unit volume of precipitate to :

$$(U)_{uyi} = \frac{m_i - 1}{m_i E_i} f_{yi}^2 + 2\varepsilon_{yi}(-f_{yi}) = \frac{m_i E_i}{m_i - 1}\lambda^2(2\alpha_i - \alpha_i^2) \quad ..(74)$$

Let a non-dimensional strain energy ratio be defined as the quotient : $\beta = (U)/(U)_u$, where β and (U) are to be furnished with suffixes characteristic for a given case and invariably :

$$(U)_u = \frac{m_i E_i}{m_i - 1}\lambda^2.$$

For the case under consideration :

$$\beta_{uyi} = \frac{(U)_{uyi}}{(U)_u} = 2\alpha_i - \alpha_i^2.$$

The above strain energy ratio β will be used for the general case of $m_i \neq m_o$ and $E_i \neq E_o$, while for the individual effect of yield in the most characteristic particular case of elastic identity, *viz.*, for $m_i = m_o$ and $E_i = E_o$, the symbol β' will be applied. It is seen that for lamellar precipitate yielding :

$$\beta'_{uyi} = \beta_{uyi} = 2\alpha_i - \alpha_i^2 \quad \ldots\ldots\ldots\ldots\ldots\ldots(75)$$

All the above equations are valid for $\alpha_i \leq 1$.

Cylindrical Precipitate of Limited Yield Resistance in Elastic Matrix

Consider a cylindrical precipitate of unit volume, having an external radius r_j and length h, to be embedded in a cylindrical matrix of internal radius r_j, external radius $r_s = \infty$, and of uniform axial length h. Note that $w_i = \pi r_j^2 h = 1$. If the radial junction stress between precipitate and external cylindrical shell is p_j, it is easily derived that the tangential stress in the cylindrical matrix at the junction r_j is $f_{toj} = -p_j$. The tangential strain of the matrix at r_j amounts to :

$$\varepsilon_{toj} = \frac{f_{toj}}{E_o} - \frac{p_j}{m_o E_o} = -\frac{m_o + 1}{m_o E_o} p_j,$$

since no axial stress develops in the matrix, *viz.*, $f_{ao} = 0$. The strain energy developing in the matrix is :

$$(2\pi r_j h)\left(-\frac{p_j}{2}\right)(r_j \varepsilon_{toj}) = \frac{m_o + 1}{m_o E_o} p_j^2 \quad \ldots\ldots\ldots(a)$$

The length of the matrix cylinder does not change under the above conditions, thus the change in volume of its cavity amounts to :

$$(2\pi r_j h)(r_j \varepsilon_{toj}) = -2\frac{m_o + 1}{m_o E_o} p_j.$$

The stresses in the yielding precipitate are at any point the same, *viz.*, $f_{ai} = p_j + f_{yi}$, $f_r = p_j$, $f_t = p_j$, are the axial, radial, and tangential principal stresses, and

$$f_{yi} = -\alpha_i \frac{m_i E_i}{m_i - 1}\lambda.$$

The change in volume of the precipitate is :

$$3\lambda + \frac{m_i - 2}{m_i E_i}(2p_j + f_{ai}) = 3\lambda + 3\frac{m_i - 2}{m_i E_i} p_j - \alpha_i \frac{m_i - 2}{m_i - 1}\lambda.$$

Equating this with the change in volume of the cavity of the matrix, the solution for p_j is obtained, *viz.* :

$$p_j = \frac{\alpha_i \dfrac{m_i - 2}{m_i - 1} - 3}{3\dfrac{m_i - 2}{m_i E_i} + 2\dfrac{m_o + 1}{m_o E_o}}\lambda \quad \ldots\ldots\ldots\ldots(76)$$

The length h of the precipitate should also remain unchanged, thus the plastic strain ε_{yia} of the precipitate, in the axial direction, is obtained from :

$$\lambda + \varepsilon_{yia} + \frac{f_{ai}}{E_i} - \frac{2p_j}{m_i E_i} = 0.$$

That is :

$$\varepsilon_{yia} = \frac{2\alpha_i\left(\dfrac{m_i - 2}{m_i E_i}\cdot\dfrac{m_i + 1}{m_i - 1} + \dfrac{m_o + 1}{m_o E_o}\cdot\dfrac{m_i}{m_i - 1}\right) - 2\dfrac{m_o + 1}{m_o E_o}}{3\dfrac{m_i - 2}{m_i E_i} + 2\dfrac{m_o + 1}{m_o E_o}}\lambda$$

$$\ldots\ldots\ldots(77)$$

being uniform all over the precipitate. The strain energy of the precipitate is, with the last term for the plastic flow :

$$\frac{m_i - 1}{m_i E_i} p_j^2 + \frac{f_{ai}^2}{2E_i} - 2\frac{p_j f_{ai}}{m_i E_i} + (\pi r_j^2)(h\varepsilon_{yia}) f_{yi}$$
$$= \frac{3}{2}\frac{m_i - 2}{m_i E_i} p_j^2 - \alpha_i\lambda\frac{m_i - 2}{m_i - 1} p_j + \frac{\alpha_i^2\lambda^2}{2}\cdot\frac{m_i^2 E_i}{(m_i - 1)^2}$$
$$- \alpha_i \frac{m_i E_i}{m_i - 1}\lambda\varepsilon_{yia} \quad ..(b)$$

The total strain energy is the sum of equations (a) and (b), that is :

$$(U)_{wyi} = \left(\frac{3}{2}\frac{m_i - 2}{m_i E_i} + \frac{m_o + 1}{m_o E_o}\right)p_j^2 - \alpha_i\lambda\frac{m_i - 2}{m_i - 1} p_j$$
$$+ \frac{\alpha_i^2\lambda^2}{2}\cdot\frac{m_i^2 E_i}{(m_i - 1)^2} - \alpha_i\lambda\frac{m_i E_i}{m_i - 1}\lambda\varepsilon_{yia} \quad ..(78)$$

The strain energy ratio, for the general case, is :

$$\beta_{wyi} = \frac{(U)_{wyi}}{(U)_u} = \frac{1}{6\dfrac{m_i - 2}{m_i - 1} + 4\dfrac{E_i}{E_o}\cdot\dfrac{m_i}{m_i - 1}\cdot\dfrac{m_o + 1}{m_o}}\cdot$$
$$\left\{9 - 2\alpha_i^2\left[\frac{(m_i - 2)(m_i + 1)}{(m_i - 1)^2} + \frac{E_i}{E_o}\frac{m_i^2}{(m_i - 1)^2}\cdot\frac{m_o + 1}{m_o}\right]\right.$$
$$\left. + 4\alpha_i\frac{E_i}{E_o}\frac{m_i}{m_i - 1}\frac{m_o + 1}{m_o}\right\} \quad ..(79)$$

and for elastic identity :

$$\beta'_{wyi} = \frac{9}{2}\frac{m - 1}{5m - 4} + 2\alpha_i\frac{m + 1}{5m - 4} - 2\alpha_i^2\frac{m + 1}{5m - 4} \quad \ldots.(80)$$

Note that, for this case, yield could not commence until :

$$\alpha_i \leq \frac{\dfrac{m_o + 1}{m_o E_o}}{\dfrac{m_i - 2}{m_i E_i}\cdot\dfrac{m_i + 1}{m_i - 1} + \dfrac{m_i}{m_i - 1}\cdot\dfrac{m_o + 1}{m_o E_o}} \quad \ldots\ldots(81)$$

In the case of elastic identity, *viz.*, for equation (80), yield may commence with $\alpha_i \leq 0{\cdot}5$. Equations (77) to (79) are valid only for values of α_i satisfying inequality (81).

Spherical Precipitate of Small Yield Stress in Elastic Matrix

A spherical precipitate is not subjected to any shear stress, consequently it is unable to yield. In spite of any small value of the yield-stress parameter α_i, the strain energy remains $(U)_v$, as given by equation (72*c*) for perfectly elastic conditions. Thus it may be written :

$$\beta_{vyi} = \beta_v \quad \ldots\ldots\ldots\ldots\ldots\ldots(82)$$
$$\beta'_{vyi} = \beta'_v = 1 \quad \ldots\ldots\ldots\ldots\ldots\ldots(83)$$

However, the above conditions predispose to instability, and such spherical precipitates should tend to change shape and then yield.

Yielding Precipitate in Elastic Matrix

Since nascent precipitation is affected by the heterogeneity of the elastic properties of precipitate and matrix, the possible influence of yield should be best characterized when elastic effects are eliminated. If elastic identity, *viz.*, $m_i = m_o = m$ and $E_i = E_o = E$, is considered, for nascent precipitates before the start of yield, $(U)_u = (U)_w = (U)_v$ or $\beta_u = \beta_w = \beta_v = 1$. When yield takes place the strain energy ratios β'_{uyi} and of course $\beta'_{vyi} = 1$, are independent of Poisson's ratio $1/m$, but that for cylindrical precipitate, β'_{wyi}, varies with the latter. To study that variation, curves were calculated for $m = 2$ and $m = 5$, and these are plotted against α_i in Fig. 22. Yield is confined to $\alpha_i \leq 0{\cdot}5$ in both cases. Note that $\beta'_{wyi} = 1$ for $0{\cdot}5 \leq \alpha_i \leq 1$. The shaded area between the two curves contains all β'_{wyi} curves possible for metals. Figure 22 contains also $\beta'_{vyi} = 1$ and β'_{uyi} plotted against α_i. It is seen that, with elastic identity, the least strain energy is associated with the least compact shape of the yielding lamella, strain energy of intermediate magnitude with the yielding cylindrical precipitate of intermediate compactness, and the greatest strain energy with the most compact shape of a spherical precipitate.

If the elastic properties of precipitate and matrix are not identical, the effect of the relative magnitude of their shear moduli should be queried. If $E_i = E_o = E$ is adhered to, but the case of $m_i < m_o$, which is equivalent to $G_i < G_o$, is considered, the strain energy ratios for cylinder and sphere both become relatively enlarged, while that for the lamella remains unchanged. In the opposite case of $m_i > m_o$, thus $G_i > G_o$, while $E_i = E_o$, the strain energy ratios for cylinder and sphere will be lower, while that for the lamella will again be unaffected. If $m_i = m_o$ is considered and $E_i < E_o$, being equivalent to $G_i < G_o$, is tried, the effect is similar to that of $m_i < m_o$; while the effect of $E_i > E_o$ is comparable with that of $m_i > m_o$. What really happens is shown in Fig. 23 for $m_i = 5$, $m_o = 2$, and $E_i/E_o = 5$, for an arrangement of the two factors that combines their effects most efficiently. For $\alpha_i \geq 0{\cdot}149$, the least strain energy is associated with fully elastic spherical precipitates, while for $\alpha_i \leq 0{\cdot}149$ yield may take place and the strain energy would become a minimum

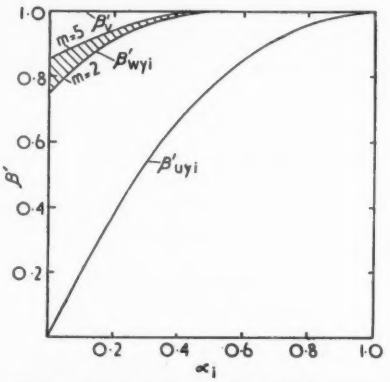

Fig. 22—Strain energy ratios for yielding precipitate having elastic identity with the matrix

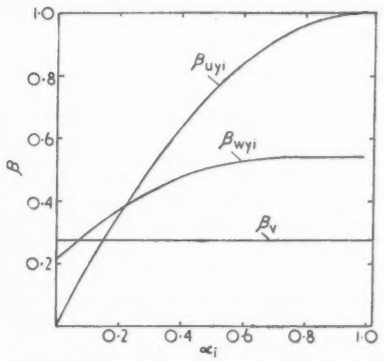

Fig. 23—Strain energy ratios for yielding precipitate for $m_i = 5$, $m_o = 2$, and $E_i/E_o = 5$

with the least-compact shape of a very thin lamella. Thus the yield curve for the cylindrical precipitate cannot have any influence on this case and it becomes apparent that, if during the process of precipitation a yielding of the precipitate becomes possible, atoms of the precipitate which are just being rejected are urged to attach themselves to the actual precipitate in such a way that the precipitate would take on the least-compact shape. Any possible variation of the individual values of Poisson's ratio and the modulus of elasticity, E, of precipitate and matrix, would not contradict the above finding.

Matrix of Limited Yield Stress Containing Elastic Spherical Precipitate

Consider a spherical precipitate of unit volume, *viz.*:

$$r_i = \frac{4}{3}\pi r_j^3 = 1.$$

where r_j is its external radius, to be embedded in an infinite spherical matrix. If the radial stress at the junction r_j of precipitate and matrix is p_j, all three principal stresses at any point of the precipitate are invariably equal to p_j. The matrix is supposed to yield between the radii r_j and r_E, the yielding volume being:

$$v_y = \frac{4}{3}\pi(r_E^3 - r_j^3);$$

thus

$$\frac{4}{3}\pi r_E^3 = 1 + v_y.$$

Since the external radius of the matrix $r_e = \infty$, it is obtained with the analysis applied on pp. 213P and 214P [IV], that the radial and tangential stresses respectively at r_E of the matrix are:

$$f_{rE} = -\frac{2}{3}f_{yo} \text{ and } f_{tE} = \frac{1}{3}f_{yo} \quad \ldots\ldots\ldots\ldots (84)$$

where:

$$f_{yo} = \alpha_o \frac{m_o E_o}{m_o - 1}\lambda$$

At any radius $r_j < r < r_E$, the radial and tangential stresses are, in terms of $v = \frac{4}{3}\pi r^3$:

$$f_r = -\frac{2}{3}f_{yo} + \frac{2}{3}f_{yo}\log_e\frac{v}{1+v_y}, \text{ and } f_t = \frac{1}{3}f_{yo} + \frac{2}{3}\log_e\frac{v}{1+v_y} \quad \ldots(85)$$

Note that for the junction at r_j, the stresses of the matrix are :

$$f_{rj} = p_j = -\frac{2}{3}f_{yo} - \frac{2}{3}f_{yo}\log_e(1+v_y)$$

$$\text{and } f_{toj} = \frac{1}{3}f_{yo} - \frac{2}{3}f_{yo}\log_e(1+v_y) \quad ..(86)$$

The yielding volume of the matrix is found from the consideration that the change in volume of the precipitate due to its volume change by 3λ and its own stresses, together with the change in volume of the yielding zone of the matrix due to the stresses of the latter, must be equal to the change in volume of the hypothetical cavity of radius r_E of the unyielding outer part of the matrix, *viz.*, from :

$$3\lambda + 3\frac{m_i-2}{m_iE_i}p_j + \int_1^{1+v_y}\frac{m_o-2}{m_oE_o}(f_r+2f_t) = 4\pi r_E^2\left[r_E\left(\frac{m_o-1}{m_oE_o}f_{tE} - \frac{f_{rE}}{m_oE_o}\right)\right] = (1+v_y)\frac{m_o+1}{m_oE_o}f_{yo} \quad .....(87)$$

it is then obtained that :

$$(1+v_y) + \frac{2}{3}\frac{m_o}{m_o-1}\left(\frac{m_i-2}{m_i}\frac{E_o}{E_i} - \frac{m_o-2}{m_o}\right)[1+\log_e(1+v_y)] = \frac{1}{\alpha_o} \quad .....(88)$$

Yield commences only if :

$$\alpha_o \leq \frac{1}{1+\frac{2}{3}\left(\frac{m_i-2}{m_iE_i}\frac{m_oE_o}{m_o-1} - \frac{m_o-2}{m_o-1}\right)} \quad .....(89)$$

The strain energy of the precipitate is :

$$\frac{3}{2}\frac{m_i-2}{m_iE_i}p_j^2 \quad .....(a)$$

The strain energy of the matrix is calculated by considering a hypothetical procedure of the volume change of the precipitate developing gradually. The first phase is that which just stresses the matrix to its yield point at r_j so that the radial and tangential stresses at r_j become, with $v_y = 0$ in equation (86) :

$$f'_{rj} = -\frac{2}{3}f_{yo}$$

and

$$f'_{toj} = \frac{1}{3}f_{yo}$$

The associated strain energy is :

$$\frac{1}{2}(-f'_{rj})4\pi r_j^2\left[r_j\left(\frac{m_o-1}{m_oE_o}f'_{toj} - \frac{f'_{rj}}{m_oE_o}\right)\right] = \frac{m_o+1}{m_oE_o}\frac{f_{yo}^2}{3} \quad ..(b)$$

Denote the combined elastic and plastic tangential strain in the matrix at r_j, during the yielding phase, by $\bar{s}_{toj}$. It may be expressed for any hypothetical value of $0 < \bar{v}_y < v_y$ as :

$$\bar{s}_{toj} = \bar{\lambda} + \frac{m_i-2}{m_iE_i}\bar{p}_j$$

where $\bar{\lambda}$ and $\bar{p}_j$ are to be expressed by the help of equations (87) and (86) respectively in terms of $\bar{v}_y$. The strain energy of the matrix during the yielding phase is :

$$\int_{\bar{v}_y=0}^{\bar{v}_y=v_y} 4\pi r_j^2(-\bar{p}_j)r_j(d\bar{s}_{toj}) = \frac{2}{3}\frac{m_o+1}{m_oE_o}f_{yo}^2\log_e(1+v_y) + 2v_y\frac{m_o-1}{m_oE_o}f_{yo}^2\log_e(1+v_y) - \frac{2}{3}\frac{m_o-2}{m_oE_o}f_{yo}^2[\log_e(1+v_y)]^2 \quad .....(c)$$

Thus the strain energy for such compound spheres amounts to the sum of (*a*), (*b*), and (*c*), *viz.*, to :

$$(U)_{vyo} = \frac{f_{yo}^2}{3}\left\{2\frac{m_i-2}{m_iE_i} + \frac{m_o+1}{m_oE_o} + \log_e(1+v_y)\left[6(1+v_y)\frac{m_o-1}{m_oE_o} + 4\left(\frac{m_i-2}{m_iE_i} - \frac{m_o-2}{m_oE_o}\right) + 2\left(\frac{m_i-2}{m_iE_i} - \frac{m_o-2}{m_oE_o}\right)\log_e(1+v_y)\right]\right\} \quad .....(90)$$

and

$$\beta_{vyo} = \frac{(U)_{vyo}}{(U)_u} = \frac{\alpha_o^2}{3}\frac{m_i-1}{m_i}\frac{m_o^2}{(m_o-1)^2}\frac{E_o}{E_i}\left\{2\frac{m_i-2}{m_i}\frac{E_o}{E_i} + \frac{m_o+1}{m_o} + \log_e(1+v_y)\left[6(1+v_y)\frac{m_o-1}{m_o} + 4\left(\frac{m_i-2}{m_i}\frac{E_o}{E_i} - \frac{m_o-2}{m_o}\right) + 2\left(\frac{m_i-2}{m_i}\frac{E_o}{E_i} - \frac{m_o-2}{m_o}\right)\log_e(1+v_y)\right]\right\} \quad ..(91)$$

If $E_i = E_o$ and $m_i = m_o$, it is obtained that :

$$\beta'_{vyo} = \alpha_o^2 + 2\alpha_o\log_e\frac{1}{\alpha_o} \quad .....(92)$$

Equations (90) to (92) are valid only for values of α_o satisfying inequality (89).

Matrix of Limited Yield Resistance Containing Elastic Cylindrical Precipitates

Consider a cylindrical precipitate of unit volume having a length h which is great relative to its external radius r_j. Note that $\pi r_j^2 h = 1$. The matrix is a cylindrical shell of length h and external radius $r_s = \infty$. If the radial junction stress between precipitate and matrix is p_j, the radial and tangential stresses in the precipitate are at any point invariably p_j. The axial stress f_{ai} for any point of the precipitate is found from the requirement that the axial length h of the precipitate should remain unchanged in spite of the volume change by 3λ. Thus from :

$$\lambda + \frac{f_{ai}}{E_i} - \frac{2p_j}{m_iE_i} = 0,$$

we have :

$$f_{ai} = -\lambda E_i + \frac{2p_j}{m_i}$$

The three principal stresses of the precipitate being f_{ai}, p_j, and p_j, and its volume being unity, its strain energy is obtained with the first half of equation (19) [II] to be :

$$\frac{\lambda^2E_i}{2} + \frac{(m_i+1)(m_i-2)}{m_i^2E_i}p_j^2 \quad .....(a)$$

For the matrix let the equations offered by Nadai[3] (*loc. cit.*, equations (50) to (52), p. 200) be considered. If the matrix yields between the radii r_j and r_E, the yielding volume is :

$$w_y = \pi h(r_E^2 - r_j^2) = \pi r_E^2 h - 1 \text{ and } \pi r_E^2 h = (1+w_y).$$

Apply the notation of $w = \pi r^2 h$, where r denotes any radius in the matrix. In the yielding zone, *viz.*, for $1 < w < (1+w_y)$, the radial, tangential, and axial stresses at the radius r, are then (in terms of volumes instead of radii) :

$$\left.\begin{aligned} f_r &= -\frac{f_{yo}}{\sqrt{3}} + \frac{f_{yo}}{\sqrt{3}}\log_e\frac{w}{1+w_y} \\ f_t &= \frac{f_{yo}}{\sqrt{3}} + \frac{f_{yo}}{\sqrt{3}}\log_e\frac{w}{1+w_y} \\ f_a &= \frac{f_{yo}}{\sqrt{3}}\log_e\frac{w}{1+w_y} \end{aligned}\right\} \quad .....(93)$$

where :

$$f_{yo} = \alpha_o \frac{m_o E_o}{m_o - 1}\lambda$$

denotes the yield stress of the matrix as before. At the boundary r_E of the yielding zone with the external elastic zone of the matrix, the stresses are :

$$f_{rE} = -\frac{f_{yo}}{\sqrt{3}};\ f_{tE} = \frac{f_{yo}}{\sqrt{3}};\ f_{aE} = 0 \quad \ldots\ldots\ldots(94)$$

The stresses of the matrix at r_j are :

$$\left.\begin{aligned} p_j = f_{roj} &= -\frac{f_{yo}}{\sqrt{3}} - \frac{f_{yo}}{\sqrt{3}}\log_e(1+w_y) \\ f_{toj} &= \frac{f_{yo}}{\sqrt{3}} - \frac{f_{yo}}{\sqrt{3}}\log_e(1+w_y) \\ f_{aoj} &= -\frac{f_{yo}}{\sqrt{3}}\log_e(1+w_y) \end{aligned}\right\} \quad \ldots\ldots(95)$$

The yielding volume w_y is found again from the requirement that the change in volume of the precipitate, together with that of the yielding zone of the matrix, should be equal to the elastic change in volume of a hypothetical cylindrical cavity of length h and radius r_E, *viz.*, of volume $(1 + w_y)$, in the external elastic zone of the matrix. This reads :

$$(2\pi r_j h) r_j(\lambda + s_{tij}) + \int_1^{1+w_y} \frac{m_o - 2}{m_o E_o}(f_r + f_t + f_a)dw = (2\pi r_E h) r_E\left(\frac{f_{tE}}{E_o} - \frac{f_{rE}}{m_o E_o}\right) \ldots(96)$$

where s_{tij} denotes the tangential strain of the precipitate, *viz.* :

$$s_{tij} = \frac{m_i - 1}{m_i E_i} p_j - \frac{f_{ai}}{m_i E_i}.$$

By carrying out the substitutions and the integration involved, it is obtained that :

$$(1 + w_y) + \frac{m_o}{5m_o - 4}\left[2\frac{(m_i+1)(m_i-2)}{m_i^2}\frac{E_o}{E_i} - 3\frac{m_o - 2}{m_o}\right][1 + \log_e(1+w_y)] = \frac{m_i + 1}{m_i}\frac{m_o - 1}{5m_o - 4}\frac{2\sqrt{3}}{\alpha_o} \ldots(97)$$

Yield commences when :

$$\alpha_o \leq \frac{\dfrac{m_i+1}{m_i}\dfrac{m_o-1}{5m_o-4}2\sqrt{3}}{1 + \dfrac{m_o}{5m_o-4}\left[2\dfrac{(m_i+1)(m_i-2)}{m_i^2}\dfrac{E_o}{E_i} - 3\dfrac{m_o-2}{m_o}\right]} \quad (98)$$

and with $E_i = E_o$ and $m_i = m_o$, when $\alpha_o \leq \sqrt{3}/2$. The strain energy of the matrix is calculated again by the hypothetical consideration of λ developing gradually from nought onward. The first stage is limited to when the stresses in the matrix at r_j just reach the yield point, *viz.* :

$$f'_{roj} = p'_j = -\frac{f_{yo}}{\sqrt{3}};\ f'_{toj} = \frac{f_{yo}}{\sqrt{3}};\ \text{and}\ f'_{aoj} = 0$$

so that the strain energy of the matrix becomes :

$$\frac{1}{2}(2\pi r_j h)(-p'_j) r_j\left(\frac{f'_{toj}}{E_o} - \frac{f'_{roj}}{m_o E_o}\right) = \frac{m_o + 1}{m_o E_o}\frac{f_{yo}}{(\sqrt{3})^2} \ldots(b)$$

During the further increase of $\bar{\lambda}$ the actual yielding volume $\bar{w}_y$ is increasing from nought to w_y. The combined elastic and plastic tangential strain of the matrix at r_j is, $\bar{s}_{toj} = \bar{\lambda} + \bar{s}_{tij}$, which is to be calculated from equation (96) in terms of $\bar{w}_y$, and next

$$d\bar{s}_{toj} = \frac{d(\bar{\lambda} + \bar{s}_{tij})}{d\bar{w}_y} d\bar{w}_y$$

is to be expressed. The strain energy of the matrix increases during this second stage by :

$$\int_{\bar{w}_y = 0}^{\bar{w}_y = w_y} (2\pi r_j h)(-\bar{p}_j) r_j d\bar{s}_{toj} = \frac{f_{yo}^2}{(\sqrt{3})^2}\log_e(1+w_y)\left[(1+w_y)\frac{5m_o - 4}{m_o E_o} - 3\frac{m_o - 2}{m_o E_o}\left(1 + \frac{1}{2}\log_e(1+w_y)\right)\right] \ldots(c)$$

where $\bar{p}_j$ has been expressed by equation (95) in terms of $\bar{w}_y$, before integration.

The strain energy for the cylindrical precipitate of unit volume is the sum of (*a*), (*b*), and (*c*), *viz.* :

$$(U)_{wyo} = \frac{\lambda^2 E_i}{2} + \frac{f_{yo}^2}{(\sqrt{3})^2}\left\{\frac{(m_i+1)(m_i-2)}{m_i^2 E_i} + \frac{m_o+1}{m_o E_o} + \log_e(1+w_y)\left[(1+w_y)\frac{5m_o-4}{m_o E_o} - 3\frac{m_o-2}{m_o E_o} + 2\frac{(m_i+1)(m_i-2)}{m_i^2 E_i} + \log_e(1+w_y)\left(\frac{(m_i+1)(m_i-2)}{m_i^2 E_i} - \frac{3}{2}\frac{m_o-2}{m_o E_o}\right)\right]\right\} \ldots(99)$$

The strain energy ratio is :

$$\beta_{wyo} = \frac{(U)_{wyo}}{(U)_u} = \frac{m_i - 1}{2m_i} + \alpha_o^2\frac{1}{(\sqrt{3})^2}\frac{m_i - 1}{m_i}\frac{m_o^2}{(m_o-1)^2}\frac{E_o}{E_i}\left\{\frac{(m_i+1)(m_i-2)}{m_i^2}\frac{E_o}{E_i} + \frac{m_o+1}{m_o} + \log_e(1+w_y)\cdot\left[(1+w_y)\frac{5m_o-4}{m_o} - 3\frac{m_o-2}{m_o} + 2\frac{(m_i+1)(m_i-2)}{m_i^2}\frac{E_o}{E_i} + \log_e(1+w_y)\left(\frac{(m_i+1)(m_i-2)}{m_i^2}\frac{E_o}{E_i} - \frac{3}{2}\frac{m_o-2}{m_o}\right)\right]\right\} \ldots\ldots\ldots(100)$$

For identical elastic constants of matrix and precipitate :

$$\beta'_{wyo} = \frac{m-1}{2m} + \frac{\alpha_o^2}{(\sqrt{3})^2}\left\{2\frac{m+1}{m} + \log_e(1+w_y)\left[4\frac{m+1}{m} + w_y\frac{5m-4}{m-1} - \frac{(m-2)^2}{2m(m-1)}\log_e(1+w_y)\right]\right\} \ldots(101)$$

Equations (99) to (101) are valid only for values of α_o satisfying inequality (98).

Matrix of Limited Yield Resistance Containing Elastic Lamellar Precipitates

Consider a square-shaped lamella of precipitate, having edges of length h and a thickness of $2r_j$ so that its volume $2h^2 r_j = 1$. Assume the edges to be rounded by the radius r_j, so that each edge may be considered as the core of a longitudinally semi-compound cylinder, the shell of which is the matrix of infinite radial extension. Thus the edges of the lamella are the cores of four semi-compound cylinders, *i.e.*, of two complete compound cylinders each having a length of h, and an external core radius r_j, *viz.*, a combined core volume for the whole lamella of $2\pi r_j^2 h = \pi r_j/h$, since $2h^2 r_j = 1$. The results of the preceding section are valid for these compound cylinders if w_y/w_i is substituted for w_y, where $w_i = \pi r_j/h$.

If the lamella of unit volume changes volume by 3λ, the change in thickness by $2r_j\lambda$ is not obstructed by the matrix if $2r_j/h$ is a very small value. However,

the change in volume by 2λ, due to the increase in length of the edges of the lamella by $h\lambda$, must be balanced by its own elastic volume change and by accommodation in the four semi-compound cylinders. Consider that the stress across the lamella is nought, and that two equal principal stresses f_i develop in its plane. The change in volume of the lamella which must be accommodated in the compound cylinders is then :

$$2\lambda + 2\frac{m_i - 1}{m_i E_i} f_i.$$

The core of the compound cylinders is furthermore subjected to the volume change by $3\lambda\pi r_j/h$, so that the combined change per unit volume is :

$$3\lambda' = \frac{h}{\pi r_j}\left(2\lambda + 3\lambda\frac{\pi r_j}{h} + 2\frac{m_i - 1}{m_i E_i} f_i\right)$$

and

$$\lambda' = \lambda + \frac{2h}{3\pi r_j}\lambda + \frac{2h}{3\pi r_j}\frac{m_i - 1}{m_i E_i} f_i.$$

The two principal stresses of the lamella in its own plane must be equal to the radial junction stress p_j of the cylinders at r_j, *i.e.*, with equation (95) :

$$f_i = p_j = -\frac{f_{yo}}{\sqrt{3}}\left[1 + \log_e\left(1 + \frac{w_y}{w_i}\right)\right]$$

considering that

$$f_{yo} = \alpha_o \frac{m_o E_o}{m_o - 1}\lambda = \alpha'_o \frac{m_o E_o}{m_o - 1}\lambda',$$

it is seen that

$$\frac{1}{\alpha'_o} = \frac{m_o E_o}{m_o - 1}\frac{\lambda'}{f_{yo}} = \frac{1}{\alpha_o} + \frac{2h}{3\pi r_j}\frac{1}{\alpha_o} - \frac{2h}{3\sqrt{3}\pi r_j}\frac{m_o E_o}{m_o - 1}\frac{m_i - 1}{m_i E_i}\left[1 + \log_e\left(1 + \frac{w_y}{w_i}\right)\right].$$

To find the value of $1 + w_y/w_i$, equation (97) is to be modified by replacing α_o by α'_o and, as has been mentioned above, by substituting w_y/w_i for w_y. Then it is to be considered that the investigation is concerned with nascent lamellar precipitates of very small thickness relative to the edge length. Thus let the modified equation (97) be multiplied by $3\pi r_j/h$, and proceed to the limit of $r_j/h = 0$. It is then found that :

$$\left[1 + \log_e\left(1 + \frac{w_y}{w_i}\right)\right] = \frac{m_i E_i}{m_i - 1}\frac{m_o - 1}{m_o E_o}\cdot\frac{\sqrt{3}}{\alpha_o}$$

and

$$f_i = -\frac{f_{yo}}{\sqrt{3}}\left[1 + \log_e\left(1 + \frac{w_y}{w_i}\right)\right] = -\frac{m_i E_i}{m_i - 1}\frac{m_o - 1}{m_o E_o}\frac{f_{yo}}{\alpha_o} = -\frac{m_i E_i}{m_i - 1}\lambda.$$

The strain energy of the lamella is :

$$(U)_{uyo} = \frac{m_i - 1}{m_i E_i} f_i^2 = \frac{m_i E_i}{m_i - 1}\lambda^2 = (U)_u \ldots\ldots (102)$$

and

$$\beta_{uyo} = \beta'_{uyo} = \beta_u = 1 \ldots\ldots\ldots\ldots (103)$$

It is consequently seen that infinitesimal lamellar precipitates will not be reduced in strain energy if only the matrix is able to yield.

Note that the above investigation of the accommodation of the two-dimensional volume change of the lamella, in the matrix, is only an approximation. The standard solution[3] (*loc. cit.*, p. 247) for plastic indentation is not applicable since the ' surface ' to be indented is not free but is cohesively jointed with the ' unstrained ' matrix on both side faces of the lamella. However, careful consideration proves that the above approximation underestimates the resistance of the matrix against indentation of the actual type. Thus the result obtained cannot be lacking ' scientific safety.' It may also be proved that the consideration of circular lamellae leads to the same end.

Yielding Matrix Containing Elastic Precipitates

As in the opposite case, just treated, assume first the precipitate and matrix to have elastic identity, *viz.*, $E_i = E_o = E$ and $m_i = m_o = m$. Curves of β'_{wyo} plotted against α_o vary with the value of m, the shaded area in Fig. 24 accounting for the range of $m = 2$ to $m = 5$. Poisson's ratio does not affect, of course, β'_{vyo} which is situated in Fig. 24 beneath the shaded area for β'_{wyo}. Since lamellar nascent precipitation is not reduced in strain energy, a horizontal line represents the case of $\beta'_u = 1$. It is seen that the least strain energy accompanies the most compact shape of a spherical precipitate, in the case of elastic identity.

Varying Poisson's ratio only, but keeping $E_i = E_o = E$, it is seen that, with $m_i > m_o$ corresponding to $G_i > G_o$, both curves β_{wyo} and β_{vyo} become elevated relative to the stationary β_u. The same result issues when $m_i = m_o = m$, but $E_i > E_o$, or when $m_i > m_o$ with $E_i > E_o$, is considered, since also these latter cases make $G_i > G_o$. In all the above cases the ordinates of the β_{vyo} curves are lower than those of the β_{wyo} curves, for one and the same values of α_o.

Greatly different modifications are observed when $G_i < G_o$ is considered, by making either $E_i = E_o = E$ with $m_i < m_o$, or $m_i = m_o = m$ with $E_i < E_o$, or $m_i < m_o$ with $E_i < E_o$. As an example the strain energy ratios are plotted against α_o in Fig. 25, using logarithmic scale for both, for $m_i = 2$, $m_o = 5$, and $E_o/E_i = 5$. If $\alpha_o > 0{\cdot}1$, $\beta_{vyo} > \beta_{wyo}$. However, this range does not actually come into play since, for $\alpha_o = 0{\cdot}1$, $\beta_{vyo} \approx \beta_{wyo} \approx 1{\cdot}65$, and with any value of α_o, elastic lamellar precipitation is associated with the strain energy ratio of $\beta_u = 1$ only. If $\alpha_o < 0{\cdot}1$, $\beta_{vyo} < \beta_{wyo}$, and least strain energy is associated with the spherical precipitate, if yield can take place.

It is to be concluded that if only the matrix can

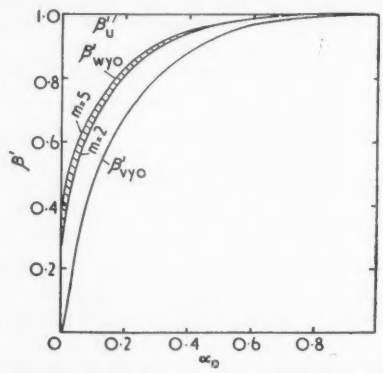

Fig. 24—Strain energy ratios for yielding matrix containing elastic precipitates, both having identical elastic properties

yield and the precipitate cannot, in order to be accompanied by least strain energy, atoms of the precipitate which are just being rejected are urged to attach themselves to the actual precipitate in such a way that the precipitate obtains the most compact shape. This rule would not be affected by any possible variation of the elastic constants of matrix and precipitate.

Some remarks should be added to the analysis of cylindrical precipitates. Nadai's equations[3] (*loc. cit.*) were derived for incompressible matrix, *viz.*, $m_o = 2$, for which case also equations for $(1 + w_y)$ and the combined plastic and elastic tangential strain, for any radius in the matrix, were given. These equations were not used above, since $m_o \gtrless 2$ was considered. However, the above results for $(1 + w_y)$ and s_{toj} become identical with those of Nadai[3] (*loc. cit.*) if $m_o = 2$ is substituted.

Objection may be raised against the consideration of the 'shear-strain energy theorem' in the analysis for cylindrical precipitation, but Nadai's equations[3] (*loc. cit.*), designed for quasi-isotropic materials, are based on it. It may be argued that precipitates are monocrystalline as well as their matrix, and that crystal plasticity appears to be controlled by maximum shear-stress rules, in general. This argument may be challenged on the ground that the above analysis was carried out using quasi-isotropic elastic constants. Nevertheless the effect of applying the maximum shear-stress theorem in the above analysis has been carefully checked. If $\sqrt{3}$ is replaced in all equations where it occurs, by the factor 2, the maximum shear-stress theorem is substituted for that of the shear-strain energy. It is then seen that all ordinates of the curves β'_{wyo} and β_{wyo} become lowered, but by a very small degree only, so that the compacting law would hold in all circumstances. The same holds also for the analysis of the cylindrical edges of the elastic lamella in soft matrix.

With the yielding lamella, or the yielding cylindrical precipitate, or the yielding spherical matrix, two principal stresses are equal so that both the above yield theorems lead to exactly the same result.

Precipitate and Matrix Both of Limited Yield Resistance

Spherical and very thin lamellar precipitates do not require any further discussion. Neither of them could make use of the twofold opportunity of yield. For the cylindrical precipitate the principal stresses are $f_{ai} = p_j + f_{yi}$, p_j, and p_j, where :

$$p_j = -\frac{f_{yo}}{\sqrt{3}}[1 + \log_e(1 + w_y)]\,;\; f_{yi} = -\alpha_i \frac{m_i E_i}{m_i - 1}\lambda\,;$$

$$\text{and } f_{yo} = \alpha_o \frac{m_o E_o}{m_o - 1}\lambda.$$

Neither the precipitate nor the matrix should change in length, thus with the notation of s_{yia} for the plastic strain of the precipitate in axial direction, we have :

$$\lambda + s_{yia} + \frac{f_{ai}}{E_i} - \frac{2p_j}{m_i E_i} = 0$$

from which

$$s_{yia} = -\lambda\left\{1 - \alpha_i \frac{m_i}{m_i - 1} - \frac{\alpha_o}{\sqrt{3}} \frac{m_i - 2}{m_i} \frac{m_o}{m_o - 1} \frac{E_o}{E_i} [1 + \log_e(1 + w_y)]\right\}.$$

The precipitate will yield if

$$\alpha_i \leq \frac{m_i - 1}{m_i} - \frac{\alpha_o}{\sqrt{3}} \frac{(m_i - 1)(m_i - 2)}{m_i^2} \frac{m_o}{m_o - 1} \frac{E_o}{E_i} [1 + \log_e(1 + w_y)] \quad \ldots(104)$$

The yielding volume of the matrix is found by the previous procedure of equating the change in volume of precipitate and yielding part of the matrix to the elastic change in volume of the hypothetical cavity of radius r_E and length h of the matrix, *viz.*, from :

$$3\lambda + \frac{m_i - 2}{m_i E_i}(2p_j + f_{ai}) + \int_1^{1 + w_y} \frac{m_o - 2}{m_o E_o}(f_r + f_t + f_a)\,dw = (2\pi r_E h) r_E \left(\frac{f_{tE}}{E_o} - \frac{f_{rE}}{m_o E_o}\right) = 2(1 + w_y)\frac{m_o + 1}{m_o E_o}\frac{f_{yo}}{\sqrt{3}}$$

where f_r, f_t, and f_a refer to equation (93), and f_{tE} and f_{rE} refer to equation (94). It is found that :

$$(1 + w_y) + \frac{3m_o}{5m_o - 4}\left(\frac{m_i - 2}{m_i}\frac{E_o}{E_i} - \frac{m_o - 2}{m_o}\right)[1 + \log_e(1 + w_y)] = \sqrt{3}\,\frac{m_o - 1}{5m_o - 4}\left(\frac{3}{\alpha_o} - \frac{\alpha_i}{\alpha_o}\frac{m_i - 2}{m_i - 1}\right) \quad \ldots(105)$$

The matrix yields if

$$\alpha_o \leq \frac{\sqrt{3}\,\frac{m_o - 1}{5m_o - 4}\left(3 - \alpha_i \frac{m_i - 2}{m_i - 1}\right)}{1 + \frac{3m_o}{5m_o - 4}\left(\frac{m_i - 2}{m_i}\frac{E_o}{E_i} - \frac{m_o - 2}{m_o}\right)} \quad \ldots(106)$$

The strain energy expression is a combination of the term (*b*) of equation (78) for the precipitate, with the terms (*b*) and (*c*) of equation (99) for the matrix, *viz.*, it reads :

$$(U)_{wyio} = \frac{3}{2}\frac{m_i - 2}{m_i E_i} p_j^2 - \alpha_i \lambda \frac{m_i - 2}{m_i - 1} p_j + \frac{\alpha_i^2 \lambda^2}{2} \frac{m_i^2 E_i}{(m_i - 1)^2} - \alpha_i \frac{m_i E_i}{m_i - 1} s_{yia} + \frac{f_{yo}^2}{(\sqrt{3})^2} \left\{\frac{m_o + 1}{m_o E_o} + \log_e(1 + w_y)\left[(1 + w_y)\frac{5m_o - 4}{m_o E_o} - 3\frac{m_o - 2}{m_o E_o} - \frac{3}{2}\frac{m_o - 2}{m_o E_o}\log_e(1 + w_y)\right]\right\} \quad \ldots(107)$$

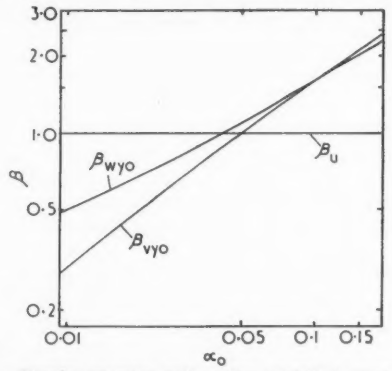

Fig. 25—Strain energy ratios for yielding matrix for $m_i = 2$, $m_o = 5$, and $E_o/E_i = 5$

The strain energy ratio is then found to be :

$$\beta_{wyio} = \frac{(U)_{wyio}}{(U)_u} = \alpha_i - \frac{\alpha_i^2}{2}\frac{m_i}{m_i - 1} + \frac{\alpha_o^2}{(\sqrt{3})^2}\frac{m_i - 1}{m_i}\frac{{m_o}^2}{(m_o - 1)^2}\frac{E_o}{E_i}\left\{\frac{3}{2}\frac{m_i - 2}{m_i}\frac{E_o}{E_i} + \frac{m_o + 1}{m_o} + \log_e(1 + w_y)\left[(1 + w_y)\frac{5m_o - 4}{m_o} + 3\left(\frac{m_i - 2}{m_i}\frac{E_o}{E_i} - \frac{m_o - 2}{m_o}\right) + \frac{3}{2}\log_e(1 + w_y)\left(\frac{m_i - 2}{m_i}\frac{E_o}{E_i} - \frac{m_o - 2}{m_o}\right)\right]\right\} \quad ..(108)$$

The latter two equations are valid only for values of α_i and α_o satisfying equations (104) and (106) respectively. With certain values of α_i and α_o, the strain energy ratio for a lamellar or spherical precipitate would become β_{uyi} or β_{vyo}, according to equations (75) or (91) respectively. By trying any possible combination of elastic constants and yield-stress parameters, it is found that either β_{uyi} or β_{vyo} is less than the co-ordinate value of β_{wyio}, in all circumstances. Hence it is seen finally that in the case of the simultaneous readiness to yield, of precipitate and matrix, either spreading or compacting will be fostered, whichever is accompanied by the least strain energy, *viz.* :

$$\left.\begin{array}{l}\text{Spreading if} \quad \beta_{uyi} < \beta_{vyo} \\ \text{Compacting if } \beta_{uyi} > \beta_{vyo}\end{array}\right\} \quad \ldots\ldots\ldots(109)$$

The latter set of inequalities is the general and universal rule as regards the possible effect of strain energy due to volume change on the shaping of precipitates when yield can take place. If $\beta_{uyi} = \beta_{vyo}$, strain energy due to volume change cannot affect the shape of precipitates during the yield phase of precipitation.

The remarks made in the preceding section as to the application of the shear-strain energy theorem may be considered also in connection with equations (104) to (108). The result is not affected by applying the maximum shear-stress theorem, *viz.*, by replacing $\sqrt{3}$, where it occurs, by 2. Inequalities (109) remain valid in any circumstance.

Criterion of Yield

The breaking-away theorem[1, 2] was associated with the idea that nascent precipitation is a perfectly elastic process. The question arises whether or not super-elasticity may exist during some initial stage of precipitation, even with fairly small yield-stress parameters, in spite of the breaking-away theorem being proved erroneous. W. L. Bragg[4] offers an analytical argument for the super-elasticity of nascent precipitation. In the light of his analysis, if the precipitate is liable to yield, it must reach a certain minimum thickness before yield can commence. If the matrix has limited yield resistance, yield does not commence unless the precipitate has obtained such a mass that the thickness of the matrix layer which would be liable to yield in accordance with given elastic constants and the yield-stress parameter, is greater than a certain minimum value. Thus the idea that any precipitation should be initially a perfectly elastic process, is to be adhered to.

The breaking-away theorem was associated, among other thoughts, with the idea of the complete relaxation of the initial strain energy. The foregoing new considerations postulate that the tessellated yield developing after the initial super-elasticity when some critical compactness of the growing precipitate has been reached, is accompanied by partial relaxation only, the degree of which depends on the actual yield-stress parameter, as shown by the following :

Strain Energy Ratios for Lamellar and Spherical Precipitation

α_i or α_o	$\beta_{uyi} = \beta'_{uyi}$	β'_{vyo}
0·9	0·99	0·9996
0·8	0·96	0·997
0·7	0·91	0·989
0·6	0·84	0·973
0·5	0·75	0·943
0·4	0·64	0·893
0·3	0·51	0·812
0·2	0·36	0·684
0·1	0·19	0·470

SHAPE OF PRECIPITATES

The foregoing figures, for spherical precipitates, for β'_{vyo}, are little affected by increasing volume fraction of precipitate rejected. However, those for the lamellar precipitate, *viz.*, for β_{uyi}, are problematical in this regard if α_i is small, since then the stress and strain energy of the matrix that develop with minute, and not infinitesimal, volume fractions of precipitate, are not negligible. Exact method of calculation cannot be offered for ' finite ' masses of lamellar precipitate. The approximate solutions, *viz.*, U for the elastic case based on equations (8) and (19) [II], and U_c for yielding lamellae based on equations (44) and (45) [IV], offer in Table XVI [IV] interesting observations. Consider the sub-rows $E2$ and $y2$ of row 5 of Table XVI [IV]. They give the strain energy per unit volume of ferrite matrix containing different amounts of lamellar graphite, the latter being incompressible, $m_g = 2$; on assuming that graphite is unyielding, $\alpha_i = 1$; and graphite has no yield resistance, $\alpha_i = 0$; respectively. Consider the columns which refer to minute quantities of graphite (' precipitate ') of 0·01 to 0·1 wt.-%, corresponding to 0·000349 to 0·00349 volume fraction of graphite. It is seen that the figures of sub-row $E2$ are, in the range under consideration, fairly proportional to the actual volume fraction, so that for large values of α_i the following consideration is justified :

(Strain energy connected with a minute lamellar precipitate of volume $\xi_i) = \xi_i(U)_{uyi}$.

As regards sub-row $y2$, with $\alpha_i = 0$ and $m_g = 2$, strain energy would not develop in the precipitate itself with any small or large volume fraction of precipitate. The figures quoted in sub-row $y2$ account for the strain energy developing solely in the matrix if the volume fraction of a 'liquid and incompressible' precipitate is not infinitesimal, but is minute. Equations (8) [II], (19) [II], (44) [IV], and (45) [IV], are, of course, fairly accurate with the small volume fractions actually concerned. The latter figures indicate that the matrix participates in the taking of strain energy, its share increasing roughly

with the second power of the volume fraction so that for small values of the yield-stress parameter α_i the following approximate relationship suggests itself :

$$\text{(Strain energy connected with a minute lamellar precipitate of volume } \xi i) = \xi i(U)_{uyi} + k\xi i^2$$

where k is a constant. However, the term $k\xi_i^2$ should be negligible for great values of α_i.

Another peculiarity in connection with lamellar precipitates is noticed when the lamella cannot yield, while the matrix has limited yield resistance. The theoretical result of no relaxation of strain energy occurring, is valid only for infinitesimal thickness of the lamella. Minute lamellae should relax. Besides this, with small values of α_o, and other conditions being suitable, buckling of the lamella might occur. One of the conditions is, of course, that the volume change of the lamella due to precipitation is an expansion.

As regards the development of shapes of nascent precipitates under conditions when strain energy due to volume change is a dominating factor of shaping, (1) lamellar shape is to be expected if $G_i < G_o$ and (2) spherical shape is to be expected if $G_i > G_o$.

If, during progressive growth in their size, yield can develop, and if strain energy due to volume change is a potent factor of shaping during the yield phase of precipitation, atoms in the process of rejection are urged to attach themselves to the actual precipitate in such a way that (*a*) individual precipitates should spread as much as possible if $\beta_{uyi} < \beta_{vyo}$, and (*b*) precipitates should undergo compaction towards the spherical shape, provided that they were not spherical before yield, if $\beta_{uyi} > \beta_{vyo}$. The latter rules are awkward to word precisely and shortly. Thus rule (*a*) will be meant when speaking of the softness of the precipitate, and rule (*b*) will be referred to when quoting the softness of the matrix.

With a lamellar nascent start, owing to $G_i < G_o$ or for some other reason, an oblate spheroid will result if the ' finish ' is according to (*a*), while a sphere or some approach to it results if (*b*)-type yield is the latter phase.

Spherical start in the nascent state will remain spherical with a yield phase (*b*) following, but should turn into a prolate spheroid, *viz.*, a cylinder in plain language, with yield conditions (*a*). The latter statement is based on the following argument. Consider, first, two possibilities of an initially spherical precipitate being turned into the least compact shape during progressing growth, *viz.*, becoming either an oblate or a prolate spheroid. Both modified units of precipitates should have the same final volume, and the polar diameter of the oblate spheroid should be comparable with the equatorial diameter of the prolate unit. Then the ratio of surface area to volume, *i.e.*, the degree of spreading, is found to be always greater with the prolate spheroid than with the oblate one. If an ellipsoid with unequal diameters has the same volume as the above spheroids, and its smallest diameter is comparable with the common diameter of the two spheroids, the degree of spreading of the former is found also to be less than that of the prolate spheroid.

The finding that the cylindrical-shaped precipitate must develop from the spherical one appears to be indisputable. The idea of initial super-elasticity is, however, not the only possible suggestion. The initial sphere may also be the consequence of surface tension or other features of the mechanics of deposition, and also of an initial yield of the matrix which is superseded by the yield of the precipitate. Such alternative possibilities will not be considered in the subsequent discussion.

The above consideration of prolate spheroids having least compactness may appear at first sight to be in contradiction to previous arguments as regards the compactness of nascent precipitates. In the latter case the lamellar shape was considered as the least compact one. However, with nascent precipitation the conditions are different. In this case the length h of a nascent cylindrical precipitate should be comparable with the diameter of a circular nascent lamellar precipitate, and their volumes should be equal, *viz.*, $\frac{1}{4}\pi d^2 h = \frac{1}{4}\pi h^2 t$, where d and t denote the diameter of the cylindrical precipitate and the thickness of the lamella respectively. The surface areas are $\frac{1}{2}\pi d^2 + \pi dh$ and $\frac{1}{2}\pi h^2 + \pi ht$ respectively. The degree of spread in shape is expressed by :

$$S_w = \frac{\frac{\pi d^2}{2} + \pi dh}{\frac{\pi d^2}{4} h} \text{ and } S_u = \frac{\frac{\pi h^2}{2} + \pi ht}{\frac{\pi h^2}{4} t}.$$

It is seen, after replacing t by d^2/h, that $S_u \geq S_w$ if $h/d \geq 1$, so that the lamellar shape proved to be the indisputable representation of least compactness, with nascent precipitation.

All the foregoing results are valid only if strains due to volume change are actually the dominating factor. Complexities may be involved if precipitation is not carried out as an uninterrupted process at constant temperature which has been implicit in this discussion. Furthermore, physical properties of matrix and precipitate, *e.g.*, yield-stress parameters, may considerably change during isothermal and uninterrupted treatment. In some such cases the ' chemical composition ' of the precipitate changes together with its hardness. This kind of change is, of course, the general rule for the matrix, and it may frequently result in considerable softening, but also in hardening to any great degree if the phenomenon of precipitation-hardening is involved. Then the yield-stress parameter may also be affected by strain hardening owing to tessellated yield.

Apparently, strains due to volume change are by no means always the controlling factor in the shaping of precipitates, so that in many cases shapes issue that are irreconcilable with the foregoing rules. The reason may frequently be recognized.

Aluminium Containing 4 to 5% of Copper

This alloy is the one which was the object of the classical studies of Preston[5] and of Calvet, Jacquet, and Guinier.[6] Both investigators arrived at the conclusion that precipitation, when carried out at temperatures not in excess of 200° C., starts with something like infinitesimal lamellae of more or less pure copper. The lattice parameters for aluminium and copper being 4·040 Å. and 3·608 Å. respectively, the strain potential due to volume change amounts

to 10·7%. With the values for E and $1/m$ of Table XI [III], it is found that $G_{Cu} = 6{\cdot}25 \times 10^6$ lb./sq. in. for copper and $G_{Al} = 3{\cdot}70 \times 10^6$ lb./sq. in. for aluminium. Since $G_{Cu} > G_{Al}$, if strains due to the volume change of the precipitate were the controlling factor as to the shape of those ' infinitesimal ' precipitates, spherical precipitates should have developed. No doubt the ' mechanics ' of deposition of the infinitesimal precipitates must have suppressed the compacting effort associated with nascent (elastic) precipitation under the actual conditions. This mechanics of deposition deserves special study.

Such infinitesimal lamellae cannot be observed nor even suspected microscopically. Their existence, size, and chemical composition are derived from observation by X-ray diffraction. It is assumed that these infinitesimal lamellae lose their identity, obviously by diffusion, in the course of ordinary precipitation, which produces microscopically visible precipitates if the actual temperature and duration of ageing make ordinary precipitation and its adequate progress possible. Under such circumstances the idea suggests itself to consider the above process as an individual type of order-hardening and not at all as precipitation proper. That would also exclude such plates from the discussion as regards the above shaping rules.

The very first ordinary precipitates are detectable by microscope after ageing at 130° C. for 7 days.[7] They appear to be spheres which are spreading into the shape of cylinders. Since $G_{Cu} > G_{Al}$, and the first phase of formation of these ordinary precipitates must take place without yield of either matrix or precipitate, a compact, *viz.*, spherical, shape should develop at some early stage. During that period, however, the yield resistance of the aluminium matrix is greatly increased by the accompanying precipitation-hardening so that yield of the precipitate should characterize the next period of precipitation. Needle-like precipitates do indeed predominate in photomicrographs corresponding to ageing temperatures of 250° C. and more, provided that over-ageing has not commenced.[6, 7]

Over-ageing is best studied with high ageing temperatures, *e.g.*, 300° C. As a point of interest, with regard to the above discussion, it may be mentioned that X-ray evidence of the ' infinitesimal ' lamellar start of precipitation is missing[6] at 300° C. Needles develop at this temperature as well as at lower ones. However, with prolonged exposure at 300° C. the aluminium-rich matrix becomes very soft owing to the gradual depletion of the copper and by the vanishing of the precipitation-hardening effect. By the end of the precipitation at 300° C. precipitates obtain a composition more or less comparable with $CuAl_2$ which is much harder than the precipitate of early periods consisting of more or less pure copper. Thus, for some final period of precipitation, the softness of the matrix becomes the dominating factor, and compacting characterizes the last stage of precipitation. Previously thin cylinders turn into stocky rods, and quite a number of fairly perfect spheres are also noticed.[6]

Alloys of Silver and Copper

Since both silver-base and copper-base alloys have very small solubility for the other metal at low temperature, but a considerable solubility at elevated temperatures, they have been favoured in investigations as regards precipitation-hardening and some other metallurgical phenomena. Calculating with the values of Table XI [III] for E and $1/m$, their shear moduli of elasticity are $G_{Cu} = 6{\cdot}25 \times 10^6$ lb./sq. in. and $G_{Ag} = 4{\cdot}01 \times 10^6$ lb./sq. in. According to the atomic spacing of 4·078 Å. for silver and 3·608 Å. for copper, the strain potential due to volume change on precipitation amounts to 11·5% with silver-base alloys and to 13% with copper-base alloys. If nascent precipitates were to develop under the shape control of purely elastic strains due to volume change, spread shape, *viz.*, lamellar silver precipitate, is bound to appear in the copper-base alloy, and compact, *viz.*, spherical copper precipitate, in the silver-base alloy.

Apparently the microscopic investigation of the shape of nascent precipitates in these alloys is fairly difficult. However, Gayler and Carrington[8] supplied perfect photomicrographic analysis for certain stages of precipitation of copper in silver-base alloys. They showed that the first microscopically visible precipitates of copper are spherical. Owing to the very great strain potential involved, yield must start with precipitates of very small individual mass and dimensions. Because of the softness of the precipitate and of the considerable precipitation-hardening of the silver matrix during elastic precipitation, the copper precipitate is bound to yield and to spread into more or less elongated cylinders. This was clearly confirmed by those experiments.[8]

In cases like the one under consideration, when the precipitate does not change by any noteworthy extent in chemical composition during the process of precipitation, its elastic properties remain constant at constant precipitation temperature. Approximately this holds also for the matrix. Then curves for the characteristic regions of yield-stress parameters may be plotted to assist the analysis of experimental observations. This is shown in Fig. 26 for both copper-base and silver-base alloys. The figure has, however, a hypothetical character since for Table XI [III] the elastic constants for room temperature were considered, instead of at the temperature of precipitation.

Consider, first, copper-base alloys with silver precipitate, and compare β_{uyi} with β_{vyo}, where i = silver and o = copper. Calculate from the equation $\beta_{uyi} = \beta_{vyo}$ points for a curve $\alpha_{Cu} = F_1(\alpha_{Ag})$, and plot α_{Cu} against α_{Ag}. If, at any instant during isothermal precipitation, the actual set of yield-stress parameters α_{Cu} and α_{Ag} corresponds to a point above the curve F_1, strain due to volume change displays a spreading effort as to the shape of silver precipitates. If the actual set of yield-stress parameters applies to a point in the plane of Fig. 26 beneath the curve F_1, the silver precipitate is subjected to compacting effort.

As regards silver-base alloys, i = copper and o = silver is to be considered when points for a curve $\alpha_{Cu} = F_2(\alpha_{Ag})$ are calculated from $\beta_{uyi} = \beta_{vyo}$. If an actual set of α_{Cu} and α_{Ag} refers to a point in the plane of Fig. 26 that is above curve F_2, compacting effort is applied to copper precipitates by strains

due to volume change, and the reverse for points beneath curve F_2.

While the two curves F_1 and F_2 are plotted in the same figure, it should emphatically be pointed out that when ageing of copper-base and silver-base alloys is considered at the same temperature, two different regions of points in the α_{Cu}–α_{Ag} plane of Fig. 26 must be co-ordinated to the alloys under consideration. With the copper-base alloy a fairly invariable value of α_{Ag} will be coupled with variable and enhanced values of α_{Cu}, while the opposite conditions hold with a silver-base alloy.

Iron and Cementite

Many processes of ferrous metallurgy involve more or less rapid changes in temperature during the treatment concerned. Though it may be suggested that the rules of the shaping effort should hold at any instant when considering the instantaneous physical constants, the case appears to be not so simple. In general, time is needed to give the atoms of the precipitate ample chance to move into such positions that the total of energies involved in the process of precipitation becomes a minimum. This should be carefully kept in mind when reading the discussion on iron and cementite and also that on iron and graphite because no reference will be made to it in the individual cases.

There is no experimental evidence which would justify assumptions that the shear moduli of ferrite, austenite, and cementite differ by any noteworthy amount over the temperature range involved. Thus shaping efforts due to volume-change strains cannot develop during the first fully elastic stage of precipitation.

When the yield phase of precipitation is considered, it will be assumed for all temperatures concerned that the softness of ferrite exceeds that of austenite, and that cementite is the hardest substance of the three. However, it would be desirable to have experimental confirmation of the latter assumption, especially for high temperatures and small rates of strain. The magnitude of the volumetric strain potentials is, of course, of great influence when opposing shaping efforts exist. Unfortunately, in some cases it appears advisable to abstain from estimating the strain potential. Strain potentials due to volume change in alloys of ferrite, austenite, and cementite, are fairly moderate. If there is doubt concerning the value of the density of one of the phases involved, caution in metallurgical considerations should be fostered by not quoting numerical figures for the strain potential.

(*a*) If ferrite is rejected from hypo-eutectoid steel above the eutectoid temperature, because of its small yield-stress parameter, spreading of the rejected ferrite should be fostered by strain energy due to volume change. In fact, photomicrographs of suitably cooled carbon steels with some 0·4–0·7% of carbon show grain-boundary ferrite, indicating that the spreading effort due to volumetric strain was given a chance of supporting the process in which excess ferrite was pushed out of the grain uniformly all over its boundary.

(*b*) The change in volume by fraction, *viz.*, the volumetric strain, which is connected with the rejection of cementite from hyper-eutectoid steel along the *ES* line of the iron/iron-carbide diagram, can be assessed[9] (*loc. cit.*, Fig. 197, p. 571) for temperatures of (700° C. ?), 800° C., 900° C., and 1000° C., as (+ 0·0242 ?), + 0·0219, + 0·0075, and − 0·0034 respectively. It is noteworthy that at a temperature of some 950° C. the density of cementite is equal to that of saturated austenite. Cementite rejected below that temperature is subjected to compression, while that separating at any higher temperature is submitted to hydrostatic tension. The (linear) strain potentials for the four above-mentioned temperatures are (− 0·0081 ?), − 0·0073, − 0·0025, and + 0·0011 respectively. Since the yield resistance of the matrix of saturated austenite is considered to be rather less than that of cementite, precipitated cementite should be exposed to compacting effort by strains due to volume change. It appears, however, that in the microstructure of such steels, after cooling at a considerable rate, a grain-boundary network of cementite having very little compactness is predominant. This is to be regarded as an undisputed victory of the mechanics of deposition over the compacting effort of volumetric strain.

(*c*) As regards pearlite formation below the eutectoid temperature in any ferrous alloy, the following consideration offers itself. If the strain potential due to volume change of cementite against ferrite is $-\lambda$, the consequential stress potential is :

$$\frac{mE}{m-1}\lambda,$$

m and E being the same for both cementite and ferrite at the temperature of pearlite formation. The stress in ferrite lamellae amounts to :

$$0{\cdot}127\,\frac{mE}{m-1}\lambda$$

and in cementite (*see* Part I) to :

$$-\,0{\cdot}873\,\frac{mE}{m-1}\lambda$$

It should be assumed that the strain potential involved is of such moderate magnitude that neither ferrite nor cementite could be caused to yield. From the point of view of shaping, pearlite formation is apparently exclusively controlled by the special

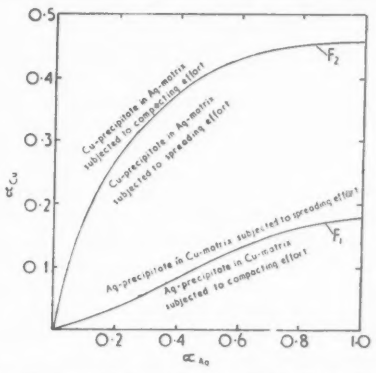

Fig. 26—Compacting and spreading effort displayed in alloys of copper and silver

mechanics of deposition, and volumetric strain cannot affect the shaping process at all.

(*d*) Upper bainite should be considered as a structure in which cementite of more or less compact shape is embedded in ferrite matrix. Depending on the actual temperature of bainite formation, and the length of exposure at that temperature, yield of ferrite might occur so that the compacting effort due to volumetric strain becomes a factor influencing the spheroidizing of cementite in upper bainite.

(*e*) While martensite formation has nothing to do with precipitation, the tempering of martensite offers itself for consideration in this regard. Tempering leads ultimately to spheroidized cementite embedded in ferrite matrix. If the matrix yields, the consequential compacting effort is a factor contributing to the spheroidization of cementite.

Iron and Graphite

While strain potentials due to volume changes accompanying precipitation in the system iron/iron-carbide are moderate, graphite separation in iron-base alloys is accompanied by extraordinarily great strains. In the range of temperature below the eutectic, the maximum specific volume of cast iron amounts[9] (*loc. cit.*, Fig. 85, p. 270) to 0·1455, as compared with the minimum specific volume of graphite of approximately 0·4444. Hence the volumetric strain accompanying graphite separation is greater than 2·0543 and the (linear) strain potential is in excess of — 0·451.

(*a*) The yield resistance of graphite is very small at room temperature. However, it may be considered that it is not much lower at temperatures up to the eutectic so that the question arises as to whether the yield resistance of the ferrous matrix at temperatures not much below the eutectic is less than that of graphite under suitable conditions. In some soft-grade grey cast iron, complicated twisted graphite lamellae are noticed. This might be an example of the buckling phenomenon, the possibility of which has already been pointed out in connection with a non-yielding lamellar precipitate embedded in a matrix of low yield resistance.

(*b*) Except for the above-mentioned possibility at temperatures close to the eutectic, when grey cast iron is cooled along the *E'S'* line of the iron–graphite equilibrium diagram, the softness of graphite ought to be the dominating feature as regards the shaping effort due to volumetric strains. Indeed the general tendency appears to be that of the formation of flakes or, in other words, the least compact shape. However, it should be considered that the shaping effort due to volumetric strain is only one among other shaping agencies. The latter remark refers to the peculiar contention of J. W. Bolton that the flakes are often more or less pencil-like rather than flat.[10] This would not be in any way contradictory to the above theory of shaping due to volumetric strains. It ought to be assumed, however, that the pencil-like shape developed from a fairly spherical one, and that the spherical shape was originated by other shaping agencies in opposition to the effort displayed by strain energy due to change in volume.

(*c*) Very interesting questions arise in connection with malleable iron. Temper carbon is of compact overall shape, furnished with a peculiar envelope having many protrusions and hence a greatly enlarged surface area. It might be argued that industrial tempering is carried out at extraordinarily low rates of change in temperature, so that cementite should become compacted, that is, more or less spheroidized. When graphite is rejected from such cementite, the shaping effect should be more or less confined to the modification of the surface of the previous cementite grain so as to make the surface-area/volume ratio great. No doubt the final surface of temper carbon would account for a successful spreading action under such conditions. However, experimental evidence should be found for such a suggestion, and furthermore there is the question as to why temper carbon rejected directly from austenite ultimately attains a compact overall shape.

The same holds for the shape of graphite rejected when cementite is decomposed during the growth of grey cast iron, or during the graphitization of steel. Neither such precipitates appear to have as little compactness as would be expected. When compared with the non-ferrous examples discussed above, the shape of graphite in ferrous alloys involves very interesting, and in some cases fairly puzzling, questions as to the nature of its development.

HARDNESS

A few considerations regarding precipitation-hardness were offered on pp. 223P to 226P of Part IV. That line of thought has since been extended.

Hardness may be defined in terms of the critical shear stress just sufficient for starting slip, or in terms of the curve of gradually increasing shear stress plotted against the shear strain, or of the average value of shear stress over a certain reference range of shear strain, or of some hardness testing method. In reading the following discussion, it should be borne in mind that hardness is used in the sense of the average shear stress over some appropriate range.

Formation of Dislocations

Experimental evidence conclusively indicates that the direction of slip in metal crystals invariably corresponds to ' lines ' of greatest atomic density. In the following, let any line of greatest atomic density be called a slip line, regardless of whether or not slip actually takes place along an individual string of atoms of this description. As regards slip planes they are either the ones of greatest atomic density or those of high atomic density.

The well-known hypothesis of dislocations depicts

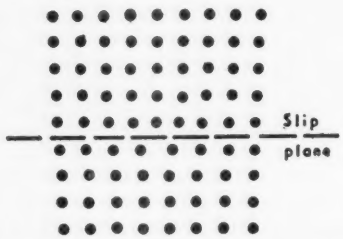

Fig. 27—A single dislocation of conventional configuration in an unstrained crystal

the string of atoms of a slip line in the neighbourhood of a dislocation such that the spacing of the atoms is reduced on one side of the slip plane, the reduction being greatest in the slip line next the slip plane ; while the spacing of the atoms is increased on the other side of the slip plane, since there is one atom less, the increase being again greatest in the slip line next the slip plane (*see* Fig. 27). Numerous considerations and analytical investigations have been offered in connection with this problem since 1934, all of which were based among other ideas on the assumption that the atoms form smooth slip lines also in the region of a dislocation.

For the closest packing of atoms, the string of atoms of a slip line may be considered as a chain, the links of which are helical springs of very great stiffness in compression, the links being coupled by universal joints. This chain is embedded in an elastic matrix. Evidently, directions of lesser and least atomic density which pass through interstices of the lattice exist in some lateral directions to the slip line of greatest atomic density. The bedding constant of the matrix is of course small in these directions. If sections of slip lines on one side of the slip plane are forced to reduce their atomic spacing, resolved along the slip line, the atoms must give way sideways and stagger into various positions around their slip lines in such a way as to consume least energy for the required shortening of the sections. These staggering movements in lateral directions are essentially two-dimensional and make a three-dimensional figure out of the previously straight chain.* For this reason no diagram is given. These displacements of the compressed sections of slip lines throw also the atoms of the surroundings out of position to a greater or lesser extent. While previous investigations considered that atoms were jumping along a smooth slip line, it is suggested that slip lines become zig-zag shaped (in two dimensions) around a dislocation.

One of the unsettled questions of the dislocation theory is how dislocations are formed. Past investigations[11] show that their formation by shear stresses which are sufficient to cause plastic shear, is impossible with smooth slip lines. The above consideration suggests a revision, keeping also the following points in mind. Two ideas appear to be indisputable, firstly that plastic shear must be a step-by-step process, and secondly that the transverse staggering displacements of atoms should be one of the essential features of this slip process. However, it may be asked whether or not the presently accepted schemes for the configuration of dislocations are the ones corresponding to least strain energy when zig-zag slip lines are considered.

If, owing to the above considered reduction in strain energy for zig-zag slip lines, it is tentatively assumed that the formation of some type of dislocations is possible exclusively by the stress causing plastic shear, then this formation process ought to be a phenomenon of instability, in a perfect lattice. The formation of dislocations should require smaller shear stresses in not perfectly ordered lattices, thus showing an analogy with the load/lateral-deflection curves of more or less imperfect columns. The above consideration is in accordance with T. A. Read's damping experiments, and connected reasoning with single crystals having lattices of varying degrees of perfection.[12]

Propagation of Dislocations

The two-dimensionally staggered atoms of the slip line cause zig-zag ' indentations ' in their bedding (matrix) which must carry out corresponding displacements roughly at right-angles to the slip line as the dislocation proceeds. With a given lattice type the resistance of the bedding to lateral displacements is, in general, least if the atomic arrangement is least pre-strained and pre-stiffened. Consequently, if the lattice has been regularly pre-stiffened by staggering atomic arrangement owing to some process previous to the shear attempt, *i.e.*, if it is ' corrugated,' the propagation of dislocations will be curbed, as was discussed superficially in Part IV for the special case of precipitation-hardening. Let the meaning of the term ' corrugated ' be extended in the subsequent discussion to include also the ' macroscopic ' corrugation of the lattice by fragmentation and even by grain boundaries, both of which display curbing effects on the propagation of dislocations as well as the atom-wise corrugation.

There is no reason to expect that the propagation of dislocations would be a perfect process in general, except for minute plastic strains in perfect single crystals. With greater than minute plastic strains, some derailing or, say, ' deranging ' of atoms must be anticipated to lead to the stoppage of the propagation of a dislocation. The deranging due to one jammed dislocation may be broken up under favourable circumstances ; however, it may also become the nucleus of a deranged cluster of atoms when the propagation of other dislocations is inhibited in the neighbourhood of the nucleus. The lattice of the deranged cluster and of its next surroundings should, of course, become corrugated and curb considerably the propagation of dislocations. While the term ' stuck dislocations ' has been in usage for many years past, it is thought that derangement means something rather different, though the meaning of the former may be extended in future usage.*

Simple Lattice

Keeping the above suggested ideas in mind, it is to be expected that least hardness would be displayed by a crystal not having a very perfect lattice so that the formation of dislocations is not retarded nor is their propagation noticeably curbed. However, the perfect lattice will carry the highest possible ' buckling load,' *i.e.*, shear stress, before dislocations are formed. Although a most easy and collapse-like propagation will characterize the accompanying slip process, the hardness as defined above would be fairly great with

* The staggering displacement of atoms caused individually by the above-considered process should be, in general, three- and not two-dimensional. However, the two lateral resolutes of these displacement vectors are significant for a qualitative discussion.

* The conception of deranged clusters of atoms offers, of course, different ideas as regards strain hardening. However, strain hardening will not be discussed in this paper.

a perfect lattice. This may be called the order-hardening phenomenon in a simple lattice. Also, this idea is supported by T. A. Read's damping experiments and connected reasoning.[12]

Mild Steel

Recently a new explanation was offered for the upper yield stress, quench ageing, strain ageing, and blue brittleness of mild steel.[13, 14] All the reasoning is based on the consideration that a ' Debye-Hückel ' atmosphere of mobile carbon atoms would segregate around existing dislocations and retard their propagation. Although it may be considered to be merely a matter of terminology, the question arises whether or not the process involved would better be designated as a special type of order hardening by carbon (and nitrogen) atoms. As to the mechanics of the hardening process, it may be alternatively interpreted as a stiffening of the lattice by the indiscriminate migration of atoms of carbon (and nitrogen) from their chance position obtained during cooling or for some other reason, to that associated with least strain energy. It should be emphasized that the arrangement of interstitial atoms so as to cause least strain energy may readily be associated with greater stiffening effects than their chance arrangement. Furthermore, it may be suggested that the stability of the atoms of carbon (and nitrogen) in their stiffening position increases with temperature, within certain limits, but can be reduced by applied shear stress. It is then unnecessary to consider this hardening process as happening only around existing dislocations. If it is again tentatively assumed that dislocations can be formed by the shear stress causing slip, the lattice stiffening by this order hardening should retard the formation of dislocations as well as their propagation.

The similarity between the stiffening effect of this order hardening and the behaviour of single crystals in Read's experiments[12] is striking. If the perfection of the lattice of a zinc crystal is impaired, slow recovery by resting at room temperature or accelerated recovery by annealing is noticed, exactly like the disappearance and return of the upper yield stress of mild steel after yield. This similarity initiated the suggestion of the term order hardening for the above phenomena in mild steel.

Precipitation-Hardening

Buckling of the lattice owing to the Preston-Guinier plates was considered in Part IV. As was pointed out above, the previous expectation[1, 2] that strains and stresses of the lattice would vanish when those plates reached a critical size by growth, is wrong. That theory would lead to the conclusion that hardening effects would disappear at that ' critical stage ' of advancing growth of precipitates. This is, of course, completely in contradiction to experimental evidence, and is not to be expected in the light of the arguments of this paper. The latter are quite compatible with experimental observations that the maximum hardness frequently develops at a stage when fragmentation of the matrix owing to yield becomes noticeable. The latter means, of course, a very intensive corrugation and stiffening of the lattice.

Theories of Hardness

Hardness phenomena displayed by metals, except those connected with chemical bonds, can all be accounted for qualitatively by the reluctance to (formation and) propagation of dislocations. While all other discussions have been focused on the slip line, in Part IV, as well as above, attempts were made to show the critical rôle of displacements in lateral directions to the slip line, and of the resistance against them. In the light of the latter considerations, the corrugation and stiffening of the lattice seems to be the predominant factor. Correlation between hardness and degree of corrugation is noticed in different cases. An apparent exception occurs in martensite : while usually the greatest increase in hardness accompanies the tetragonal lattice-distortion in order hardening of solid solutions, practically no reduction in hardness is noticed when tetragonal martensite is turned into cubic by tempering. It could, perhaps, be assumed that the ' cubic corrugation ' of martensite would be just as effective as the ' tetragonal corrugation ' in producing hardness. However, the reduced density of tetragonal martensite may also be considered as a critical factor in this case, keeping in mind that this should promote the formation and propagation of dislocations. This reasoning means that not only the degree of corrugation and stiffening, but also the ' atomic porosity,' is an essential factor as regards hardness. Fortunately, the degree of lattice corrugation and stiffening has not been mathematically defined yet, so that its use as a criterion of hardness has the advantage of restriction to qualitative arguments. Its correlation with hardness may be a fairly complicated problem also in cases other than martensite.

Keeping in mind that tessellated yield should, in general, increase the corrugation of the matrix on the one hand, and on the other that with $\alpha_o = 0$ precipitation-hardening should not develop at all, it becomes evident that the objections raised in Part IV against the attempt to correlate precipitation-hardness with the strain potential due to volume change[15] were justified. It appears that precipitation hardness ought to be correlated with some quantity which depends on the strain potential and yield-stress parameters involved.

Satisfactory correlation is obtained when strain potential due to ' atomic ' misfit and hardness of plain solid solutions are considered,[16*] since in such cases yield does not complicate the problem.

SUPPLEMENTARY NOTES

Problems connected with tessellated stresses and their analysis are fairly complex. Nevertheless, it was necessary owing to war-time conditions to compress Parts I to IV into the least possible space. Methods of analysis were gradually developed in the course of the years during which that series of papers was written, and various results and inferences were superseded. Unfortunately, the revisions in the earlier work which were automatically and unmistakably necessitated by the results of the latter studies, for the sake of brevity, were not always mentioned.

* *See* also other literature quoted in reference 16.

It has therefore been of great value that the contents of Parts I to IV were recently reviewed by F. R. N. Nabarro[17] (*loc. cit.*, p. 61). The review contained various criticisms most of which were materially withdrawn by Nabarro in the discussion[17] (*loc. cit.*, pp. 426–431). E. Orowan made some critical reference[17] (*loc. cit.*, pp. 47 and 424) to Parts I to IV, both directly and indirectly. J. A. Wheeler[17] (*loc. cit.*, p. 407) and others considered points which were topics of Parts I to IV. For the sake of establishing facts it appears necessary to reconsider here a few questions which could not be settled in the discussion[17] (*loc. cit.*, pp. 426–431) on account of limitations imposed by time and space, and to add some remarks regarding the points with which Orowan and Wheeler were concerned. Also a few new considerations will be offered.

Stress and Strain Energy Associated with Martensite Formation

Martensite is subjected, in general, to strains and stresses due to (*a*) shearing forces between martensite and austenite, and between cohesive martensite plates, and (*b*) the difference in density between austenite and martensite. Orowan merely discussed, qualitatively, stresses of type (*a*), and also their influence on the progress of transformation and the retention of austenite.

In Part I the stresses (*b*) due to the difference in density were exclusively analysed because they were readily accessible to fairly exact statistical calculation. While that analysis was carried out without taking the possible yield of austenite (and martensite) into numerical consideration, such refinements could easily be made with the help of the methods developed in Part IV and in this paper.

It is mentioned on page 147P of Part II that the stress analysis for lamellar tessellation in Part I, is deficient since the stress concentration at the edges of the lamellae remains neglected, and that the latter is to be approximated by the analysis of a compound cylinder. In the light of this idea, the consideration of compound cylinders and Figs. 5 [I] and 6 [I] appear to be a very rational approach as regards stresses (*b*), now even as earlier. The sharpest scrutiny could not do more than suggest the modification of the abscissa of Figs. 5 [I] and 6 [I] since the volume fraction of austenite is not simply interconnected with that of the ' edge-cylinders.'

This refutes the second criticism of Wheeler. His first objection is concerned with strains and stresses in the initial stages of transformation when martensite plates are surrounded by austenite. However, page 157P of Part II should be recalled. For the whole process of transformation, not strains and stresses, but the strain energies involved, are of direct importance. For the strain energy associated with change in density only, equation 21 [II] holds, since the elastic constants of austenite and martensite are comparable. The stress analysis criticized by Wheeler leads in the simplest and most direct way to equation 21 [II]. The one suggested by him would apparently be a combined analysis of stresses (*a*) and (*b*). It seems, however, that the rational approach to the ' complete solution ' should not be based on Wheeler's suggestion, but on the separate consideration of stresses and strain energies (*a*) and (*b*). This would not affect the analysis of stresses (*b*) which was criticized by Wheeler.

Retained Austenite

Retained austenite possesses some ' potential energy ' for its transformation, but not enough. Suggestions were made by Orowan and by Wheeler (*loc. cit.*) concerning the reasons why austenite should be retained. Wheeler's ideas were refuted by Orowan so that nothing further need be said about them. Orowan's suggestion is based on the shearing forces developing between martensite and austenite. The resulting shear stresses in austenite should (*a*) stabilize it against further transformation, and also (*b*) promote the formation of martensite on some other shear plane. There seems here to be some contradiction. Unless this deficiency is remedied, the discussion of the suggestion should be suspended, except for the remark that reflections in terms of strain energy were necessary.

Let the idea pointed out on page 191P of Part I and worded more precisely on page 227P of Part IV, be remembered. It was suggested that the ratio of surface area to mass of an individual particle of retained austenite is the critical factor. The boundary between retained austenite and the martensite matrix was considered to require great modifications as regards misfit atoms, if particles of retained austenite should (fully or partially) be transformed. The ' potential energy ' of austenite that is available for transformation is proportional to its mass. The energy required for the rearrangement (and the ' formation ') of misfit atoms is, however, proportional to the surface area of the individual particle of retained austenite.

Furthermore, provided that austenite yields and becomes fragmented, so acquiring a ' greatly increased ' grain-boundary area, there must result an increase in the reluctance of austenite to transform, and more ' potential energy ' is necessitated. Yield and fragmentation should originate from the boundary of austenite with martensite and penetrate inwards into the mass of austenite. Thus the overall effect of this procedure, per unit mass of austenite, should again be influenced by the ratio of surface to mass of an individual particle of austenite.

The lattice of martensite is greatly corrugated, and the lattice of austenite is forced to conform to that corrugation over a certain depth of its boundary layer. It may also be suggested that this might necessitate increase in reluctance to transform, without yield and fragmentation of austenite being taken into consideration. Since no dislocation, but a different type of shear process is involved in martensite formation, this suggestion appears to be problematic for the time being. In any case, the average magnitude of any type of lattice corrugation in austenite must be considerably affected by the ratio of its surface area to its mass.

If the ratio of surface to mass is a critical factor, not only the retention of austenite at room temperature and its reduction by sub-zero treatment find explanation, but also the limited extent of transformation if austenite is subjected to isothermal hardening at a temperature above atmospheric.

The two new arguments just discussed may be modified so that the average degree of yield and fragmentation or the degree of the mere corrugation is related to the mass of martensite which 'supplies' the energy, and to the mass of austenite in which the special features for the increase in reluctance to transformation are developed. Thus the criterion of increase in reluctance due to these new arguments would be referred to the ratio of the actual proportions of martensite and austenite. This latter criterion is, of course, more or less comparable with the other one, *viz.*, the ratio of surface area to mass of individual particles of retained austenite. The area/mass criterion would readily account for the fact that large single masses of retained austenite have not been noticed, whilst the alternative criterion would not exclude them. However, it is possible that both criteria are significant and are acting simultaneously.

Volume Fraction of Retained Austenite

It was suggested on page 191P of Part I that, with increasing rate of cooling, (1) the size of individual particles of retained austenite diminishes, but (2) the number of particles of retained austenite increases, so that at some intermediate rate of cooling there is a maximum in the volume fraction of retained austenite. The intermediate rate is, of course, meant to be greater than the critical one. Both in Nabarro's review[17] (*loc. cit.*, p. 61) and in his discussion[17] (*loc. cit.*, pp. 426–431), arguments were advanced that, owing to the above suggestions (1) and (2), the minimum of retained austenite should be expected (at some intermediate rate of cooling). It was expressed there in some way that formally both suggestions are possible, the actual outcome being controlled, say, by the shape of the influence curves (1) and (2).

Epstein[18] shows curves of percentage of retained austenite plotted against cooling rate for three plain carbon steels. A steel with 0·41% of carbon has a critical cooling rate of 875° C./sec., at which about 9·7% of austenite is retained. This point occupies more or less exactly the apex of the curve, which is, however, drawn by some intuition in this region since the neighbouring experimental cooling rates are 440° C./sec. and 1250° C./sec. Another steel with 0·82% of carbon has a critical cooling rate of 440° C./sec., accompanied by 15·3% of retained austenite. The retained austenite reaches a maximum of 16% with a cooling velocity of 750° C./sec. Both points are experimental ones. The third steel, with 1·30% of carbon, shows the point of critical cooling rate to occupy the apex of a curve which is drawn again with much intuition in this range.

On considering that a section of the graph has been omitted, it is noticed that none of the three curves shows a minimum at some intermediate rate of cooling. The result is one maximum and no minimum, which is thus in opposition to the suggestion in Nabarro's review and discussion. Nevertheless, a mistake was made in Part I by alleging that maxima were in general found by experiment.

Random Aggregates of Compound Units

The analysis of numerous problems was approached by using a certain type of simple consideration, leading to equations (8) [II], (9) [II], (10) [II], and (14) [II], and to equations (44) [IV]. Owing to the aim at extreme brevity, the arguments produced for the formulation of these equations must have created a very doubtful impression. Equations (8) [II] were singled out and criticized in Nabarro's review; while the supplementary arguments put forward for their justification in the discussion were accepted by him, some misunderstanding occurred on that occasion also. It was not the present author's intention to characterize them as suffering from any degree of 'absurdity'; they are no doubt deficient to some extent, like any other approximations. Agreement could not be reached on their unquestionable acceptability as approximations. However, what else is left until a better approximation has been produced? Furthermore, it should be noted that it was equations (8) [II] which made possible an exact (not approximate) proof that the strain energy of a lamellar precipitate is not at all affected by the 'breaking away' of the precipitate from the matrix, in contradistinction to previous beliefs.[1, 2]

Zinc, Cadmium, and Tin

It was suggested on page 200P of Part III that the metals zinc, cadmium, and tin should develop some degree of preferred orientation on heating and cooling, owing to recrystallization. This recrystallization was expected to be a consequence of yield due to the excessive strains caused by heating or cooling in random aggregates of such metallic crystals, and it was supposed to take place during the heating and cooling cycles (and also after their completion). The development of some degree of preferred orientation would make possible a reduction in residual strain energy due to the change in temperature, and would thus be a consequence of nature's effort towards least energy.

In Nabarro's review (*loc. cit.*) objection was raised against the above idea with the argument that "recrystallization is an isothermal process." In the discussion there was a request for the source of the statement just quoted and in the reply reference was made to Ewing and Rosenhain[19] as the ones who first observed "isothermal recrystallization at the temperature of straining" in lead. However, Ewing and Rosenhain happened to report[19] that lead strained exclusively at room temperature was successfully recrystallized, not only at room temperature, but also at 100°, 150°, 250°, and 300° C. It should furthermore be added that Calvet, Trillat, and Paic[20] cold-rolled pure aluminium at room temperature and successfully recrystallized it subsequently at 0° C. Readers who believe that recrystallization is an isothermal process should therefore be told that, on the contrary, it takes place, provided that nuclei exist, at any constant or permanently varying temperature between certain upper and lower limits. The lower limit of temperature of recrystallization is room temperature for zinc and cadmium, and below room temperature for tin.[21]

It should be concluded that the above idea of isothermal recrystallization is untenable and would not justify the epithet "unsound"[17] (*loc. cit.*, p. 61) in connection with the suggestion made in Part III. In his reply[17] (*loc. cit.*, p. 431), some further remarks

were made by Nabarro in this context which are apparently the result of some misunderstanding because heating or cooling of specimens free from external forces was considered in Part III.

Redistribution of Tessellated Stresses

The redistribution of tessellated stresses of structural type and of those due to the cooling of random aggregates of crystals having directionally variable thermal expansion, was discussed on pages 204P to 206P of Part III. Cold working was considered as a suitable means to redistribute such stresses. It was pointed out that individual crystals of a random aggregate should be supposed to undergo different degrees of plastic deformation and to sustain different magnitudes of stress during cold working. On removing the ' external forces ' causing cold working, such crystals should be furnished with some degree of misfit so that an individual type of tessellated stress develops between crystals. In Part III it was suggested that these stresses " may be . . . of a less order of magnitude than were those which had been reduced by cold working." This idea was criticized in Nabarro's review. The objections were based on a pair of ' theories,' *viz.* :

(*a*) " That the stresses required for cold working are of the same order of magnitude as the internal stresses originally present," and

(*b*) " That the internal stresses produced by cold working are of the same order of magnitude as the applied stresses."

Discussion took place on this point. A further argument against the above theories has been developed since the closure of the discussion. Let the term cold working in this case refer only to those slip processes which are the result of the propagation of dislocations. Force, and hence energy, is primarily necessary for the propagation of dislocations in any solid crystal of any high degree of freedom from internal stresses, and of any great perfection of the atomic arrangement. The more perfect is the actual process of slip by the propagation of dislocations, the less is the fraction of the energy required for the propagation of dislocations that can remain as residual strain energy owing to cold working. Though the force and energy necessary for the propagation of dislocations may be fairly small with a soft metal crystal, they are of a ' greatly ' different order of magnitude from zero, which represents the internal stresses in the unstrained crystal. Thus theory (*a*) fails also in this fundamental ' experiment,' together with theory (*b*) which is stated in the discussion to be the consequence of (*a*).*

Surface Layer

The behaviour of the surface layer has been carefully studied and discussed from different aspects in Parts I to IV. Its deficiency with regard to fatigue was suggested on pp. 158P and 159P of Part II and on p. 203P of Part III. This idea was challenged in the review[17] (*loc. cit.*, p. 61), but was accepted owing to the arguments put forward in the discussion[17] (*loc. cit.*, p. 431).†

* Work in hand will enable further progress to be made with regard to strain hardening.

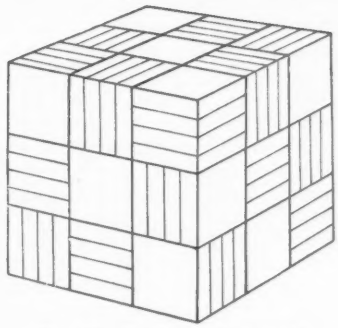

Fig. 28—Basic random unit

On the present occasion, let a further point be quoted in connection with the surface layer and fatigue, *viz.*, the phenomenon of ' size-deficiency.' It is well known that test specimens of increasing cross-sectional dimensions, made of the same material, display fatigue strengths falling with increasing size. Consider that the periodic force applied to the specimen contains also upper harmonics of the fundamental, and assume that their intensity is proportional either to the cross-sectional area or to the section modulus of the specimen. Since the area of the surface layer is proportional to the square root of the cross-sectional area and the cubic root of the section modulus, the surface layer of large specimens must become relatively overstrained in comparison with that of smaller ones, if certain upper harmonics concentrate their straining effort to the surface layer. To expect such a contention does not seem to be quite unreasonable on analogy with the result of the analytical investigation of earth-tremor.[22] This idea would, of course, lead to the conclusion that size-deficiency should not be noticed if the ' effective ' upper harmonics are filtered out, which could be arranged in fatigue experiments ; or if the surface layer is appropriately reinforced, *e.g.*, by case hardening or nitriding.

Tessellated Stresses

Figure 28 shows that certain interfaces of the idealized random aggregate of units of tessellation look like a tiled wall, each tile being associated with an individual stress system. The visualization of this picture suggested the introduction of terms like ' mosaic ' or ' tessellated.' The former has, however, been prejudiced by usage for minute fractions of crystals. Though tile-pattern stress systems are the fundamental feature even of the atomic structure, such stresses are centred, in the case of concrete, around large pieces of stone having edges up to a

† It is noted on p. 269 of A. Kochendörfer's book, " Plastische Eigenschaften von Kristallen und Metallischen Werkstoffen " (Berlin, 1941 : Springer), that this problem has been experimentally investigated by R. Glocker and H. Hasenmaier, *V.D.I.*, 1940, vol. 84, p. 825. They found that the surface layer of plain carbon steel undergoes plastic strain at stresses which are between one-half and two-thirds of the stress at which the bulk of the cross-section yields. This is a welcome confirmation of the suggestion (*see* Parts II and III) that the surface layer possesses an inherent weakness as regards yield stress.

few inches. Thus preference was given to the term 'tessellated' in spite of its being a "somewhat sibilant expression"[17] (*loc. cit.*, p. 62).

E. Orowan (*loc. cit.*, p. 42) objected explicitly to this term. He found it too definite and not applicable to many types of stresses in his group 'textural stresses.' It does not appear, however, that any of the examples discussed by Orowan could not be accommodated under the heading 'tessellated stresses.' Furthermore, the term 'textural' has also been prejudiced by past usage in connection with the preferred orientation of crystals. It is thought in addition that scientific English gives preference to definite terms. This is demonstrated by Professor H. O'Neill's suggestion 'chequered'[17] (*loc. cit.*, p. 399). That adjective would indeed be admirable if also its figurative meaning were to be taken into consideration, seeing that reflections on these stresses tempt the investigator very strongly into qualitative speculations, whereas their analysis too often contradicts those expectations and the connected 'theories.' However, 'tessellated' being identical in concrete meaning with 'chequered,' let the former be retained for the time being.

Tessellated stresses, as such, were censured also for their being complicated, and the desire was expressed for such 'without tears'[17] (*loc. cit.*, p. 399). It should, however, be realized, as some comfort to the reader, that the analysis of tessellated stresses is very much farther from being 'without tears' than is the study of the publication of the results. Owing to Nabarro's recent review, its discussion, and the above contributions, the idea suggests itself that a possible method of overcoming the difficulties connected with their study may be the discussion of the papers with their author himself. With the kind permission of The Iron and Steel Institute, discussion on the present paper may be extended to the previous Parts I to IV.

Acknowledgments

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B. I. S. R. A.

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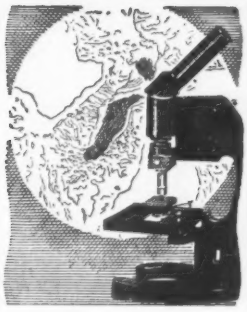

The Breakdown of Austenite below the M_s Temperature

By F. C. Thompson, D.Met., and M. D. Jepson, M.Sc., Ph.D.

SYNOPSIS

After demonstrating that there is no reason to doubt that austenite in a plain carbon eutectoid steel can decompose isothermally below the M_s temperature, it is suggested that the decomposition during the actual quenching operation and that due to the isothermal transformation should be considered quite independently, and so shown in the isothermal transformation diagram.

The results of dilatometric investigations show that, so far as the second factor is concerned, decomposition becomes progressively slower as the temperature is lowered, and that there is no sign of a knee in the curve. The nature of the transformation product in this range of temperature is considered, the conclusion reached being that it is martensite in the first instance, which tempers progressively as the time is increased. This leads to an acicular, bainitic product, which is of secondary origin.

Determinations of the maximum rate of breakdown, $R_{max.}$, suggest that this is related to the temperature by an equation of the type $R_{max.} = Ke^{k\theta}$. There appears to be a distinct break in the curve of temperature against this maximum rate at the M_s temperature.

The characteristic $\alpha\beta$ inversion of the martensite can be observed dilatometrically immediately after quenching, but it disappears as the isothermal treatment is prolonged and the martensite transforms.

The mechanism of the austenite–martensite transformation during isothermal treatment is discussed, and no reason is found to believe that it is not due to mechanical stress, just like that resulting from the quenching itself. Finally, suggestions are thrown out as to the origin of this stress, and magnetic results are recorded which reveal changes in the austenite during the induction period.

THE TRANSFORMATION OF AUSTENITE

THE nature, the mode of formation, and even the very existence, of the isothermal transformation product in the martensite range have been the subjects of much discussion for a number of years, and the position cannot even yet be said to be completely satisfactory.

Within a few years of the publication of Davenport and Bain's[1] paper, two main schools of thought had come into existence concerning the reactions represented by that part of the **S** curve below the knee. The **S** curve itself indicates that austenite transforms isothermally to martensite at a rapidly increasing rate as the temperature is reduced. In opposition to this, Carpenter and Robertson,[2] for example, claimed that martensite starts to form during cooling as soon as the steel reaches a certain temperature, now known as the M_s point, which is characteristic of the steel. Cooling below the M_s temperature results in a further formation of martensite, but when cooling ceases the transformation is arrested and, provided that the temperature is not so low that the transformation has already been completed, will remain incomplete so long as the temperature remains constant.

Greninger and Troiano[3] attempted to decide between these diametrically opposed views. Their experiments showed that transformation took place both during the cooling and also under isothermal conditions, the conclusions being :

(1) Martensite begins to form as soon as the specimen temperature reaches the upper limit of the martensite range. The transformation continues as long as cooling proceeds, the degree of transformation depending on the temperature down to which the material is cooled.

(2) If cooling is halted, then the formation of martensite stops for a time, which is longer the lower the temperature, and then restarts isothermally at a

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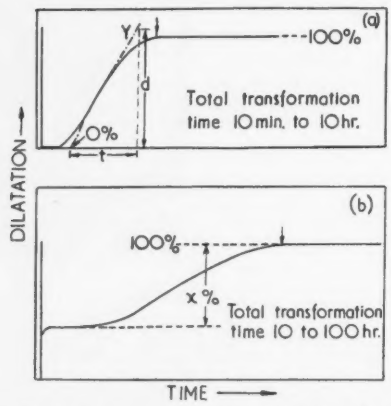


Fig. 1—Typical dilatation/time curves taken from dilatometer records. (*a*) Temperatures above M_s, (*b*) temperatures below M_s. Construction lines in (*a*) indicate method of calculating the maximum rate of dilatation. (Drawings are not to scale)

very slow rate. This slow isothermal transformation will proceed to completion if sufficient time is allowed.

It should be noted that in their view the rate of isothermal transformation continues to decrease in the lower temperature range and not to increase, as the **S** curve suggests.

The present authors' dilatation/time curves (using the Stanton dilatometer[4]) are in full agreement with the conclusions drawn from the metallographic work by Greninger and Troiano. The curves obtained fall into two distinct groups, shown in Fig. 1, comprising (*a*) *all* transformations at temperatures above M_s, and (*b*) *all* transformations below M_s.

The first group (Fig. 1*a*) showed a short period of induction immediately following the quench, and then a fairly rapid increase in length, gradually slowing down until dilatation eventually ceased. The second group (below M_s, Fig. 1*b*) showed curves which consisted of three distinct parts : firstly, a rapid increase in length for the first minute or so ; secondly, a long period of induction ; and finally, a slow dilatation taking many hours, in some cases weeks, to come to completion.

The initial rapid dilatation observed on quenching below M_s can be explained as being due to the formation of martensite during the cooling itself. Once the temperature has fallen below M_s, two effects will exist in opposition, namely, the contraction due to cooling *A*, and an expansion *B* due to the breakdown of the austenite. The curves for each of these may be expected to be of the forms indicated in Fig. 2, giving a net result *C*, very similar to that obtained experimentally.

These results are entirely consistent with those of Greninger and Troiano, but there is one possible explanation of an apparently ' isothermal ' transformation which would seem worthy of consideration. As has already been shown,[5] small fluctuations in temperature can cause a relatively rapid change from austenite to martensite, apparently due to short periods of cooling. Such fluctuations are almost certain to occur in any so-called isothermal quenching bath, for example, owing to the intermittent action of the thermostat. It is possible, therefore, that the ' isothermal ' breakdown of the austenite may, in reality, be due to flashes of transformation occurring at intervals when the temperature of the bath is falling slightly. In other words, truly isothermal transformation below M_s may not, after all, exist.

To control the temperature of a quenching bath with sufficient accuracy to eliminate this effect would be a matter of considerable difficulty, but it was felt that until it was reasonably certain that isothermal transformation did take place, further discussion of the whole problem was useless.

If the premise that cooling is essential to the formation of martensite is accepted, then clearly, since constant temperature is inhibitive, a slowly rising temperature should tend even more to prevent transformation. Further, by the imposition of a steadily rising temperature the possibility of periods of cooling can be eliminated, and if transformation is found to take place under these conditions it would provide very strong grounds for the belief that under isothermal conditions breakdown can and does occur.

To examine this possibility, the temperature of the quenching bath was arranged to rise at rates of 1°, 2°, or 3° C./hr. Continuous records of the bath temperature covering the whole period of treatment were obtained, and in no case was there any indication of cooling.

The specimens for use in the dilatometer were 1 in. long and 0·16 in. in dia. ; those required solely for metallographic examination were shorter so that a number could be mounted, polished, and etched together. To prevent decarburization, all the specimens were nickel-plated before treatment, and were heated in a stream of dry nitrogen. The metallographic specimens were heated and quenched in batches and then withdrawn singly from the bath and quenched. To distinguish between the products of ' isothermal ' transformation, if any, and that produced on quenching, specimens transformed wholly within the martensitic range were at first tempered at 300° C. for 30 sec. before being quenched in cold water, but subsequent work shows that this procedure was unnecessary.

Concerning the dilatometric experiments, it was realized that any increase in length recorded by the

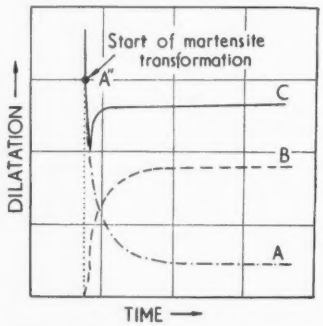


Fig. 2—Formation of the first part of the dilatation/time curve in the martensitic range

dilatometer would be the sum of the thermal expansion due to the rising temperature, and the volume change due to the transformation. To separate these it was necessary to have at least an approximate knowledge of the thermal expansion, and the change in length of a standard specimen of the same steel in the annealed state was measured under conditions as near as possible to those employed in the experiments. The coefficient of expansion calculated from these results is not, of course, identical with that of the steel in an austenite–martensite condition, but it is near enough for the present purpose. The expansion values obtained in this way were subtracted from the total measured dilatation, thus giving that part of the dilatation due solely to the transformation.

Samples were quenched at various temperatures in the range 140–350° C., most runs below 200° C. being repeated at each of the three rates of temperature rise. Typical curves are shown in Fig. 3, but are plotted on a much reduced time scale, since some of the records taken at low temperatures were well over 6 ft. long. It will be observed that in Fig. 3*a* the thermal expansion, due to less than 1° C. rise in temperature, was too small to have a measurable effect and has been ignored. In all cases the transformation/time curves were of normal shape, and the effects of variations in the rate of temperature rise were slight.

In Fig. 4 the dilatometer records of a series of quenches at temperatures ranging from 150° to 320° C. are plotted on a logarithmic scale. To avoid confusing the diagram, only the 'isothermal' transformation subsequent to the period of induction has been recorded. For this reason the apparent dilatation within the martensite range decreases as the temperature

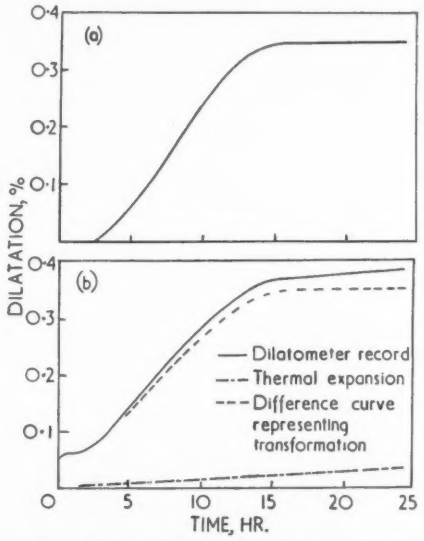


Fig. 3—Transformation with the temperature slowly rising. Typical dilatation/time curves. (*a*) Above M_S ; (*b*) below M_S, showing the subtraction of the thermal expansion from the measured dilatation to obtain the transformation curve. (Drawings are not to scale)

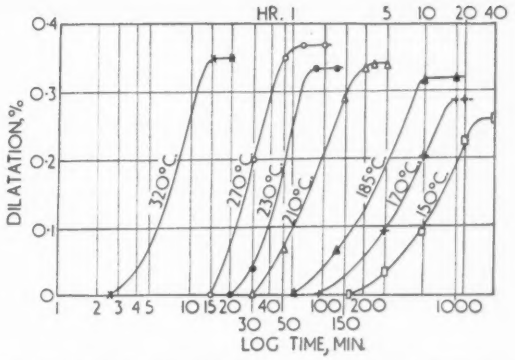


Fig. 4—Rising temperature. Dilatation/time curves obtained at various temperatures. Temperature rise 2° C./hr. Cooling transformation is not shown

is lowered, owing to a progressively higher degree of transformation during the quenching operation.

In addition to the dilatometric results given above, groups of specimens were treated and prepared for metallographic examination. The specimens treated at 160°, 200°, and 250° C. (initial bath temperature) all showed that transformation proceeded under the rising temperature, a typical set being shown in Fig. 9, for samples taken from a series quenched from 800° to 200° C., the bath temperature rising 2° C./hr.

From this evidence, it is therefore quite clear that transformation of austenite *does* take place at temperatures below M_S, even when the temperature is slowly raised. It can, therefore, be concluded that truly isothermal transformation does take place within the martensitic range.

THE ISOTHERMAL TRANSFORMATION DIAGRAM

It is, perhaps, at this point that the most convenient way of recording isothermal data should be considered. The diagram typified by the **S** curve has done yeoman service but clearly requires considerable modification in the light of later work.

Up to the present time two main types of modified isothermal transformation diagrams have been

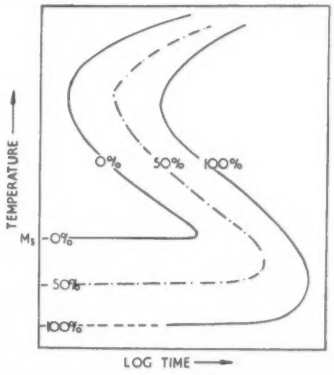


Fig. 5—Isothermal transformation diagram for high-speed steel (Gordon, Cohen, and Rose)

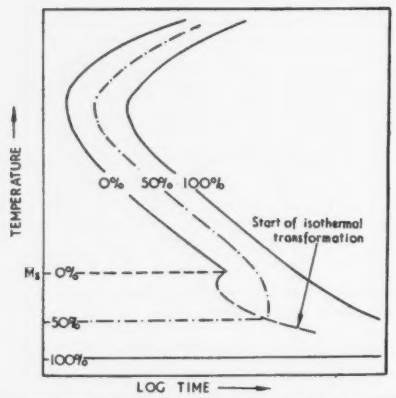


Fig. 6—Isothermal transformation diagram for 1·1% C, 1% Cr steel (Payson and Klein)

proposed—the type used by Cohen[6] and his co-workers, Fig. 5, and the type proposed by Payson and Klein,[7] Fig. 6, being examples. Both of these satisfy the requirements to some extent; the M_s temperature is indicated, and the dependence of the initial transformation to martensite during cooling, on the degree of that cooling, is represented. It is suggested that it would be a further improvement if the essential difference in the two modes of transformation, *i.e.*, during cooling, and then as a result of the isothermal breakdown, were made still clearer.

The most satisfactory method may be to use two separate diagrams, as can be done very conveniently by the superposition on the isothermal transformation diagram of an additional diagram showing the extent of the breakdown during cooling. Figure 7 is drawn in this way for steel *A*, a 0·84% carbon, 0·55% manganese steel.[5] The **C** curve represents only the transformation/time/temperature relationships for the isothermal transformations; earlier changes which occur during the quenching are ignored. The square diagram, on the other hand, shows the amount of transformation which had occurred during cooling to a particular temperature, *i.e.*, it is a transformation/temperature diagram only. It also indicates the percentage of the total transformation which remains to be completed isothermally if the cooling is arrested.

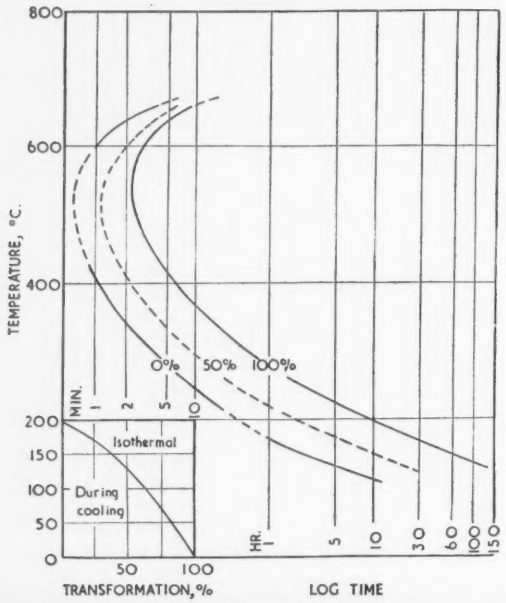


Fig. 7—Isothermal transformation diagram for steel *A*

The square diagram can be built up in any of the usual ways, in this case being based on dilatometric measurements. Over a large number of experiments it was found that, above 200° C., the extent of the dilatation was almost independent of temperature, the percentage dilatation varying between 0·35 and 0·39% with most of the values close to the average, 0·37%. Assuming that the total volume change below M_s is not very different from that at higher temperatures, it was possible to estimate from the available data the proportion of the transformation taking place during the cooling. To do this, the measured dilatation during the isothermal part of the transformation, *i.e.*, x%, Fig 1*b*, was subtracted from the average total dilatation given above, the difference being expressed as a percentage; thus, if x is 0·25%, the unchanged austenite is

$$\frac{0{\cdot}25}{0{\cdot}37} \times 100 = 67\%,$$

whence 33% decomposed during the quenching.

This method of calculation is by no means accurate, since for example, the volume change involved in the austenite–martensite inversion is greater than that to any other product. In addition, the experimental data employed were accurate only to about $\pm$ 5%, which is probably the order of the error in the final result. However, the accuracy obtained is probably sufficient to show the general trend of the quenching transformation.

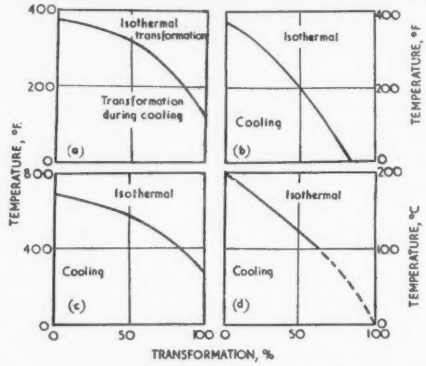
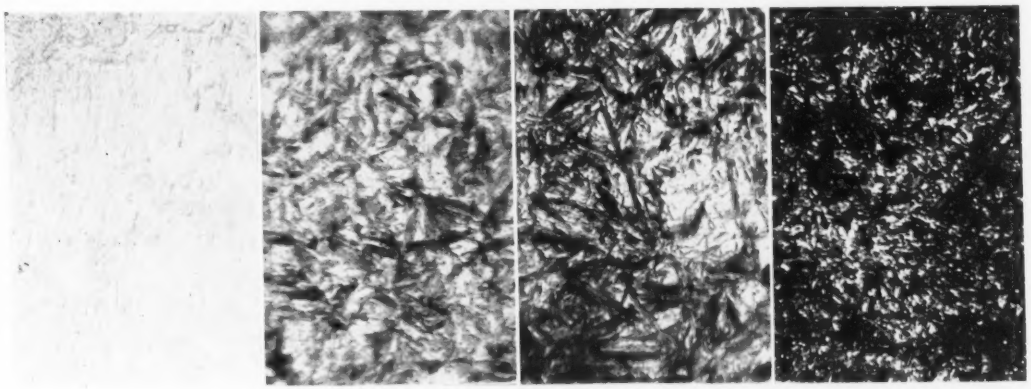


Fig. 8.—'Square' diagrams taken from transformation diagrams of (*a*) Payson and Klein, (*b*) Gordon, Cohen, and Rose, (*c*) Finn, Cook, and Fellows, and (*d*) taken from Fig. 7

For comparison purposes the results of certain other workers have been plotted in the same way, Fig. 8, and similar curves have been obtained, even with steels of widely differing compositions.



(*a*) 1 min. (*b*) 1 hr. (*c*) 2 hr. (*d*) 6 hr.

Fig. 9 — Transformation with rising temperature. Samples quenched from 800° to 200° C., the bath temperature rising 2° C. hr. Transformation times as shown. Etched with 4% nital × 1000

(*a*) 10 min. (*b*) 2 hr. 40 min. (*c*) 5 hr. 20 min.

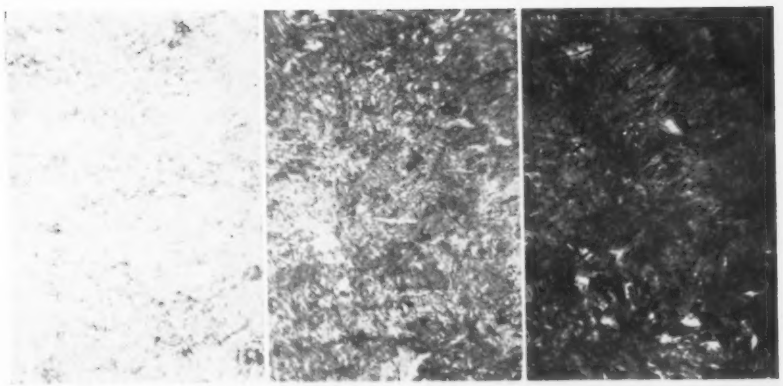
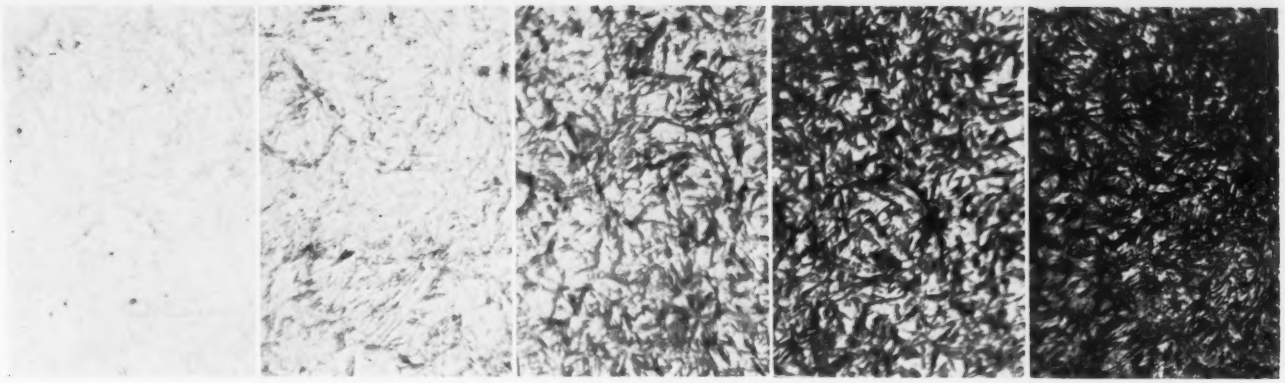


Fig. 10—Effect of tempering martensite at low temperatures. Specimens water-quenched and tempered at 150° C., for various times. Etched with 4% nital for 6 sec. × 400



(*a*) 1 hr. (*b*) 4 hr. (*c*) 6 hr. (*d*) 18 hr. (*e*) 25 hr.

Fig. 11 — Low-temperature transformation. Samples heated to 800° and quenched at 150° C. Transformation times as shown. Etched with 4% nital for 6 sec. (a)–(f) × 350, (g) × 1400

(*f*) 30 hr. (*g*) 30 hr.

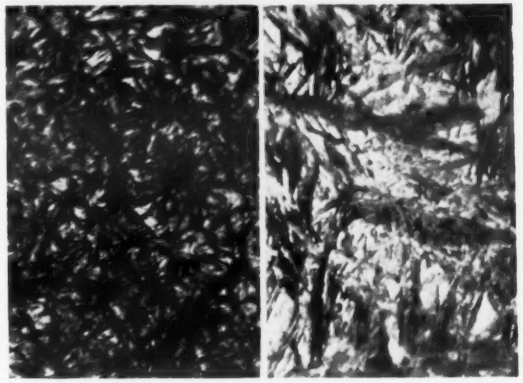


Fig. 12—Twinned structure of original austenite revealed by needles produced by full transformation × 350

NATURE OF THE ISOTHERMAL PRODUCT BELOW THE M_s TEMPERATURE

Mehl[8] has shown that the nuclear growth of pearlite has surrendered to the mathematical approach. High-temperature bainite can also be explained with only slight modifications ; but the formation of low-temperature bainite is less easily explicable on such a basis, whilst martensite is usually considered to be the product of slip or shear within the crystal, initiating the reorientation from the face-centred to the body-centred cubic lattice.

Thompson and Millington[9] were perhaps the first to suggest that martensite might be formed by some such mechanism : " Under the stress to which the crystals of austenite are subjected during quenching, the ' easy glide ' movement* . . . is initiated. Normally in a close-packed cubic metal this would . . . cause the production of twinned material . . . In the present case, however, the fact that they are produced from an unstable component will, in general, cause the atoms . . . to *reorientate themselves entirely* in a stable form, and thus become α-iron. At this stage we should have bars of α-iron, still containing its carbon in supersaturated solid solution, traversing the unaltered austenite."

Regarding the origin and nature of the isothermal product below M_s, Davenport and Bain, in their original paper, designated the product martensite, but more recently the suggestion has been put forward[10, 11] that the isothermal product is more likely to be of the nature of low-temperature bainite. The evidence cited in support of this theory is of two kinds. In the first place, the product is relatively dark-etching. Secondly, it is pointed out that there is no clear break, or marked change in direction, of the **C** curve of the isothermal transformation diagram at M_s, which might be expected if an entirely new product started to form below this critical value.

Considering the etching criterion first, it may be pointed out that etching effects, by themselves, though clearly of importance, provide uncertain evidence. So much depends on the matrix in which a phase is embedded and on its state of dispersion. One example which may be cited is the case of the structure of an αβ-brass, in which the β-phase normally etches dark and the α-phase light. The same material, however, mildly quenched from, say, 800° C.—at which temperature it is all β—shows a feathery precipitation of α at the crystal boundaries. Etching, as before, in ferric chloride acidified with hydrochloric acid, now darkens the α-crystallites, leaving the β-phase unattached. It may also be pointed out that in the form of diagram suggested by Payson and Klein, a break at the M_s temperature is definitely suggested. This point will be discussed later.

There is good evidence for the belief that martensite, produced directly by quenching, is an unstable phase even at temperatures well below 100° C. This has been shown by Campbell,[12] who measured the resistivity of a quenched 0·95% carbon steel after tempering it at that temperature for increasing periods of time. As a result, the resistivity fell from 36·32 microhms/c.c. in the quenched state, to 27·31 microhms, after only 12 hr. at 100° C. Even at room temperature the resistivity was shown to fall in two years from 35·85 to 32·58 microhms. The decrease at 100° C., 9·01 microhms, at any rate, is far too high to be explicable as a mere stress-relief effect, and there can be little doubt that a definite change of some kind is proceeding.

Recent X-ray work by Kurdjumov and Lyssak[13] has led to the conclusion that, at temperatures of the order of 150° C., only a part of the carbon can remain in solution if such tempering is prolonged. This decomposition may not, of course, go to completion ; it seems, in fact, much more likely that it will tend to some steady state, the amount of carbon remaining in solution depending on the temperature. Carpenter has, for example, published the microstructure of a hardened steel tool some 4000 years old, showing microscopically typical martensite.

Since such tempering effects produce a marked difference in the etching characteristics, it may be possible to explain the etching properties of the isothermally produced ' martensite ' on similar lines. To examine this further, a eutectoid plain carbon steel was quenched from 800° C. into cold water and tempered in oil at 150° C. for various lengths of time. The specimens were then carefully mounted in one group and were prepared together for microscopic examination. This ensured that all the specimens received identical treatment during polishing and etching, and so were strictly comparable. The photomicrographs in Fig. 10 show quite distinctly the progressive change which takes place.

In addition to the microscopic examination, hardness measurements were carried out, and, as will be seen from the following results, the tempering treatment effected a substantial reduction :

Variation in Hardness of Fully Quenched Eutectoid Steel on Tempering for Various Times at 150° C.

Tempering Time	Vickers Pyramid Diamond Hardness Number
5 min.	905
10 min.	893
50 min.	876
1 hr. 20 min.	862
2 hr. 40 min.	852
5 hr. 20 min.	845
48 hr.	741

These results, taken in conjunction with the resistivity changes noted by Campbell, and the X-ray results of Kurdjumov and Lyssak, suggest strongly that the tempering of martensite at quite low temperatures will cause precipitation of at least some of the carbide. Since, then, martensite does appear to break down at temperatures well below M_s, it was desirable that an attempt should be made to determine whether an analogous effect was to be observed during isothermal transformation.

The same eutectoid steel as before was used in the form of nickel-plated specimens, 0·5 in. long and 0·16 in. dia. The specimens, mounted on thin steel wire carriers, were heated to 800° C. in a stream of dry nitrogen and were then quenched in a molten solder bath maintained at 150° C. From previous

* Shear on the {111} face of the austenite.

dilatometric work it was expected that approximately 50% of the austenite would transform during the quench, and this was subsequently confirmed by the microstructure. On maintaining the samples in the bath for periods of up to 4 hr., the induction period, the martensite needles were observed to etch progressively darker, but no indication of any increase in their number was observed. Continued treatment at the same temperature resulted in further decomposition of the austenite, and more needles, which could easily be distinguished from the original crop, began to form. After, say, 10 hr. the majority (*i.e.*, the original, quenching, needles) etched even more darkly than before, but they were now admixed with others which etched much more lightly, presumably part of the new formation at the isothermal temperature. As the time of the treatment was prolonged, these in their turn darkened, and more light needles were produced to darken again progressively.

This process may be followed in the typical series of microstructures shown in Fig. 11. It should be noted that the preparation and etching of all the specimens was most carefully standardized, as was the preparation of the photomicrographs, so as to avoid all possibility of spurious results. After the specimens had been allowed to transform isothermally for the required time, they were quenched in cold water, so that any untransformed austenite was converted into martensite of the normal light-etching variety.

In addition to the metallographic examination, hardness measurements were made, the hardness decreasing progressively as the transformation time increased. The hardness numbers of the specimens shown in Fig. 11 are given below:

Change in Hardness with Progress of Transformation at 150° C.

Transformation Time, hr.	Vickers Pyramid Diamond Hardness Number
1	873
4	859
6	852
18	800
25	775
30	760
44	736

By the kindness of the English Steel Corporation, Ltd., it was possible to attempt to measure the hardness of individual needles of this low-temperature product on a microhardness apparatus. The structures were too fine to permit of very accurate measurements being carried out, but the results indicated clearly that the darker needles were distinctly softer than the lighter ones, and that both were softer than the background martensite present after the final quenching in the partially transformed specimens.

The ranges of hardness covered in these results, for which, as has been noted, no great accuracy can be claimed, were:

Very dark needles	650–700 D.P.N.
Light needles	750–800 D.P.N.
Martensitic matrix	850–900 D.P.N.

All these results, metallographic and mechanical, are consistent with the view that the immediate product of decomposition in the isothermal bath is not in any way different from the martensite formed on quenching. It can be concluded, therefore, and this again is consistent with the resistivity and X-ray results mentioned, that the austenite does not pass directly to the final product of decomposition, but through an intermediate 'phase,' the characteristics of which suggest that it is martensite.

It does not appear to be stretching the evidence too far to suggest that the acicular forms of bainite represent a structure which was originally martensitic and which has subsequently broken down to the final product. The authors find from their own work no support for the view that they represent any primary product of transformation.

VARIATION OF MAXIMUM RATE OF TRANSFORMATION OF AUSTENITE WITH TEMPERATURE

Further light is thrown on to the subject under discussion if the rate of isothermal transformation of austenite, and in particular, the variation of its maximum rate with temperature, is examined. Over a number of years a large number of dilatation/time curves for steel *A* have been accumulated, from which it is possible to measure the maximum rate of decomposition. To do this the tangent to the curve, at the point of maximum rate of dilatation, $R_{max.}$, was drawn and its slope was calculated in terms of percentage dilatation per minute, as indicated by the construction in Fig. 1*a*.

On plotting $R_{max.}$ against the temperature of transformation, the curve shown in Fig. 13 was obtained. This consists of two straight lines of somewhat different slopes, and shows a very distinct gap around 200° C., the M_s temperature.

In a discussion of these results, Dr. L. B. Pfeil pointed out that the slope of the dilatometer curve would be affected by the amount of unchanged austenite still available for transformation. The expansion during transformation depends on the amount of austenite left, a quantity—at the start of isothermal

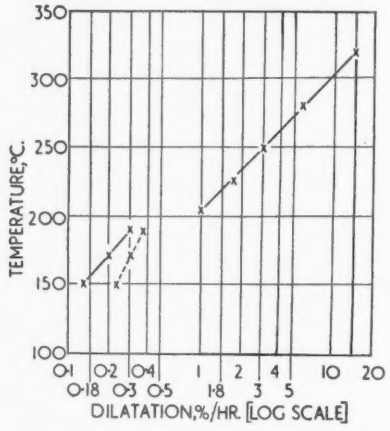


Fig. 13—Variation in maximum rate of transformation with temperature. Calculated from dilatometric results

transformation at any temperature below M_s—which can be obtained from the square diagram in Fig. 7. As a rough indication, therefore, of the actual behaviour of the residual austenite, the rate of dilatation, calculated as above, was multiplied by a factor of two when half the austenite had broken down during quenching, by three when this fractional transformation had been two-thirds, and so on. When this correction was applied, as is indicated by the corrected line, shown dashed in Fig. 13, the break at 200° C. is still perfectly clear, and the difference of slope of the curves is even greater.

The relationship suggested by these results, *viz.* :

$$\frac{dE}{dt}\ (\text{max.}) = Ke^{k\theta}$$

does not seem to bear any obvious connection with any normal type of chemical reaction, and is, it is suggested, indicative of a complex transformation due, not to a physico-chemical reaction, but to some state of mechanical stress.

The evidence, then, provides grounds for the belief that the breakdown of austenite, above and below the M_s temperature respectively, is different, and may well depend, in part at least, on quite different physico-chemical factors.

There are, therefore, three main lines of argument, none of which lends support to the view that below the M_s temperature the initial breakdown product of the austenite is bainitic :

(1) There is a marked change in the form of the dilatation/time curve at M_s, Fig. 1.

(2) The progressive change of the etching characteristic during isothermal treatment, which can only be explained by the formation of a primary, unstable ' phase ' which subsequently tempers.

(3) There is a marked directional change in the curve relating the maximum rate of transformation and temperature at M_s, which may be taken to imply that the processes involved in the formation of the breakdown products above and below this temperature are in some way different.

The only fact remaining to support the bainite theory is the continuity of the **C** curve (Fig. 7) through the M_s temperature. The transformations in this range are, however, slow, and the errors are fairly large—a factor masked by the use of a logarithmic scale. It will be noted, however, that if the curve, Fig. 13, above M_s is extrapolated downwards, it would appear to cut, or merge into, the low-temperature curve, a result in excellent accord with the break in the dotted curve shown in Payson and Klein's diagram (Fig. 6).

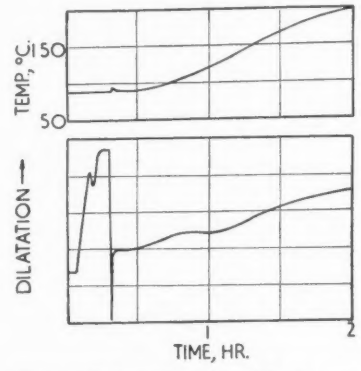


Fig. 14—Tempering curve showing an arrest at the martensite inversion. Quenched from 800° C. at 90° C. and transformed for 2 min. at 90° C. before tempering

THE $\alpha\beta$ MARTENSITE INVERSION

In an attempt to settle the question beyond all reasonable doubt, it was decided to try to identify the breakdown constituent below M_s by observation of the $\alpha\beta$ martensite inversion.

The eutectoid steel was quenched from 800° C. in a bath at 90° C., as a result of which some 70% of the austenite decomposed to martensite during cooling. On *immediately* reheating this specimen, an inversion point was clearly marked on the dilatometer record, Fig. 14, in the temperature range 110–120° C.

Similar quenched samples were then allowed to remain in the bath at 90° C. for periods of 30 and 200 hr. respectively. The former had not quite completed the induction period, while the latter had undergone about half of the full transformation of the residual austenite. On reheating, using the maximum sensitivity of which the apparatus was capable, neither specimen showed any sign of the $\alpha\beta$ change.

So far as the specimen treated for 30 hr. was concerned, this is further proof of the instability of martensite at these low temperatures. The absence of the inversion point in the specimen which had remained in the isothermal bath for 200 hr. suggested, at first sight, that no martensite was present. This argument loses its cogency, however, when it is realized that only a small amount of untempered martensite would be present at any one time, and that only an apparatus of extreme sensitivity could be expected to detect the very small change in volume involved.

The evidence available, then, seems to justify certain tentative conclusions, and to suggest that the most probable process and product of transformation below M_s may be as follows :

When the austenitic steel is quenched to a temperature within the martensitic range, a certain amount of transformation takes place, the proportion decomposed depending on the temperature down to which cooling is carried. The product of this breakdown is martensite. On holding the material at constant temperature, the martensite already formed begins to temper. At the close of the induction period, the length of which increases as the isothermal temperature is lower,* further transformation to martensite is initiated and continues until all the austenite is decomposed. While this decomposition of the austenite is taking place, tempering continues, affecting all the martensite present. When all the austenite has broken down, the final structure will then consist of martensite which has tempered to varying degrees. The structure still shows its acicular nature, since the martensitic needles, having tempered to different extents, have varying etching properties, and the final product is a tempered martensite, *i.e.*, acicular bainite.

* All our results agree on this point, *see* Fig. 7.

MECHANISM OF ISOTHERMAL TRANSFORMATION BELOW M_s

So far only the real existence and the nature of the isothermal product have been considered, and no attempt has been made to define the causes producing the change. If, as the authors consider probable, the product of isothermal transformation below M_s is, initially, martensite, its formation should be due to a mechanism similar to that resulting in the formation of the same constitution during quenching, that is, shear initiated by stress.

The reasons which lead the authors to this conclusion are:

(1) So far as the most careful microscopic examination can show, the distribution and microscopic characteristics of the needles are the same in both cases.

(2) Decomposition under isothermal conditions is accelerated by the imposition of external stress (particularly tension).[5]

(3) Under such stresses of a magnitude just sufficient to initiate transformation, the needles in those crystals of austenite which have shown signs of breakdown are disposed roughly at 45° to the direction of the applied stress.[5]

(4) Where few needles have been produced by the stress, they are strikingly suggestive of mechanical twins.[5]

(5) The needles bear a definite relationship to the crystalline structure of the parent austenite. This is well shown in Fig. 12, in which the twinned structure of the original austenite has been revealed by the needles produced by full transformation.

On the basis of these facts it would appear to be justified to conclude that the martensite due to isothermal decomposition, just as is that resulting from the quenching, is the result of mechanical stresses producing shear on definite crystallographic planes. Passing on to consider the origin of such stresses, it is clear that under isothermal conditions, purely thermal stresses cannot be operative and some other source must exist. In order to produce internal stress some volume change must take place within the specimen, and the change must also be dependent on the thermal conditions, since temperature has a pronounced effect on the rate of breakdown. These internal strains must be small, since a sensitive dilatometer cannot detect them, perhaps owing to their being highly localized, and may build up slowly since their effect may become noticeable only after a relatively long induction period.

Such internal strains may be visualized as being produced in a number of ways. They must clearly originate either in the martensite formed during the quenching or in the residual austenite, and both are possible. The tempering of the martensite has been fully discussed above, and the effect of the precipitation of carbide during the process will be to cause a decrease of volume which will impose a stress on the austenitic matrix. When the stress becomes sufficiently great to initiate shear in those crystals most suitably orientated, a further formation of martensite will result. This will temper in its turn, the stresses again rising until a further flash of transformation takes place.

Changes may also be occurring within the austenite. If, for instance, it be true that the carbon atoms tend to occupy certain preferential positions in the martensite lattice,[14] migration of the carbon atoms must occur, and diffusion can take place, even with atoms as small as those of carbon, only at the expense of marked local strain of the lattice.

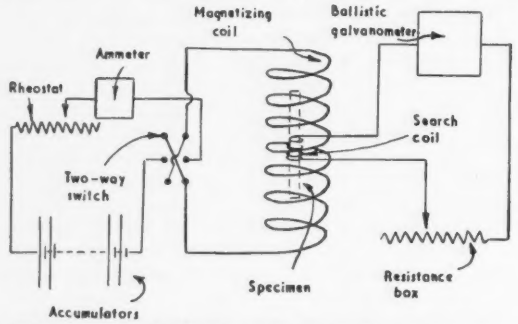


Fig. 15—Circuit diagram of magnetic apparatus

MAGNETIC CHANGES DURING THE INDUCTION PERIOD

That the induction period is not one of complete quiescence may be demonstrated in a qualitative manner by examining the magnetic characteristics.

In this preliminary work only comparative results have, as yet, been obtained, and the simple circuit used is adequately illustrated in Fig. 15. The magnetizing coil consisted of some 7000 turns of copper wire wound on a $\frac{3}{8}$-in. dia. silica tube between asbestos end plates, and gave a maximum field strength of 2500 gauss. The search coil consisted of 40 turns connected through a resistance adjusted to give critical damping of the galvanometer.

The coils were immersed in the oil-quenching bath and the whole was maintained at a temperature of 150° C. The specimen was quenched, slipped into the coil, the magnetizing current was switched on, and the galvanometer was allowed to settle down. The throw of the galvanometer, on reversing the current, was noted at intervals, two reversals being

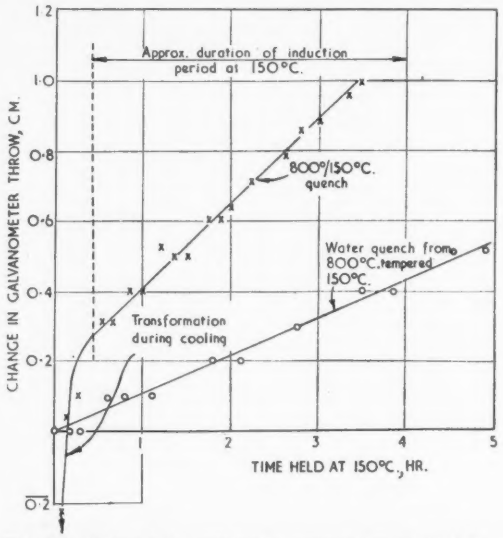


Fig. 16—Change in magnetic properties at 150° C.

made at each stage to give throws in opposite directions. To ensure that there was no drift in the apparatus, caused by the elevated temperature, measurements were made on an annealed specimen at the same temperature. Readings taken over a period of 5 hr. showed no significant change.

The two most interesting experiments were as follows :

(1) The specimen was quenched from 800° C. into the oil at 150° C. and the change in permeability was followed over a period of 4 hr.—approximately the duration of the induction period at this temperature for the steel used. A steady increase of permeability occurred, amounting in all to about 10%, Fig. 16.

(2) A second specimen was first water-quenched from 800° C. to give a more or less fully martensitic structure. After about 5 hr. in the oil bath at 150° C. the ballistic throw had again increased, but by only about 3%. These results have been repeated a number of times under slightly different conditions and amply confirmed.

The difference between the two samples, then, can be taken to be real, and it may be inferred that there is a much greater change in the magnetic properties of a steel quenched to 150° C. and held at that temperature, than can be explained by the tempering of the martensite formed during the quench, and that some other change, *i.e.*, in the residual austenite, must be taking place at the same time.

CONCLUSIONS

(1) Since doubts have been expressed as to the possibility of the truly isothermal decomposition of austenite below the M_s temperature, it has been shown that this transformation may be induced even when the temperature is slowly *rising*.

(2) It is then shown that there is a marked difference between the dilatation curves obtained above and below that temperature.

(3) The best manner of representing the results of isothermal breakdown is considered, and the suggestion is made that the change during the quenching should be measured, and represented, quite independently of that due to the isothermal treatment. Apart from this, the results are in very good accord with the type of diagram proposed by Payson and Klein and with the micrographic work of Greninger and Troiano.

(4) The nature of the isothermal product is discussed and the grounds for the belief that this is originally martensite are given. Tempering of this martensite may proceed side-by-side with further breakdown, leading to an end-product, bainitic in character, with an acicular structure. The hardness of the needles has been roughly determined, and it is shown that this decreases progressively as the time in the isothermal bath is increased.

(5) The maximum rate of decomposition, $R_{max.}$, has been determined as a function of temperature, and it is suggested that the two are related by an equation of the type :

$$R_{max.} = Ke^{k\theta}$$

Since this does not seem to have a very obvious connection with any normal type of chemical reaction, it is suggested that mechanical stress is a dominating factor.

(6) The reasons are given which have lead the authors to believe that the mechanism of the formation of martensite during isothermal decomposition is essentially identical with that resulting in martensitization during quenching. In both cases it appears to be the result of stress inducing movement on some definite crystallographic plane, but the authors have, as yet, no evidence that this plane is the same in the isothermal product.

(7) Such internal stresses may be set up either in the martensite through volume changes due to tempering, or by changes in the austenite. Diffusion, for instance, might provide the explanation in the latter case, where, as magnetic results demonstrate, the austenite is by no means quiescent during the induction period.

Acknowledgment

The authors wish to express their acknowledgment of the financial assistance given by the British Iron and Steel Research Association, which rendered it possible to carry out this work.

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Department of Scientific and Industrial Research

"*Report for the Year* 1947–48," published recently by His Majesty's Stationery Office, reviews ten years of development in industrial science and the part played by the Department in peace and war. The Command Number of the Publication is 7761, and the price 2*s*.

Vocational Training in the Metal Trades

A report entitled "*Vocational Training and Promotion in the Metal Trades*," issued by the International Labour Office, urges comprehensive training facilities in the metal industry. The report was prepared for the third session of the I.L.O.'s Metal Trades Committee, held at Geneva in November.

Split Thermocouples

By A. Searby, A.Met., A.I.M.

THE measurement of liquid steel temperatures is now a routine operation in steelworks, but there is no means yet of providing a continuous record of steel bath temperatures by the use of thermocouples. A method which has been suggested involved the leading of the thermocouple elements through the furnace lining into the metal at some distance apart, the molten charge thus completing the circuit and acting as the 'hot junction.' The advantages would seem to be: (i) the absence of thermal lag; (ii) the extremely small area of contact with the metal; and (iii) new thermocouple wire is brought into contact with the bath as the furnace lining wears back, thus eliminating the errors associated with contaminated wires. A trial of the method in a side-blown converter had previously given negative results, and it was with the intention of obtaining further information and completing the trial that the following laboratory investigations were made, using a 10-lb. high-frequency furnace.

EXPERIMENTAL PROCEDURE

Two probes were made, each consisting of a 12-in. length of small-bore silica tubing, one carrying a 0·5-mm. dia. platinum wire, and the other a 0·5-mm. dia. platinum–13% rhodium wire. The wires were cemented into position with high-grade alumina cement, and were left protruding approximately 1 in. The bare ends were then plunged into a bath of molten pig iron to a depth of $\frac{3}{4}$ in., and held 2 in. apart. The metal surface was clean and free from slag. A reading was obtained instantly on a manual potentiometer, and remained constant for 2 min., after which the thermocouple was removed. The reading of 1258° C. was confirmed by an ordinary type of immersion pyrometer. The 'split' thermocouple was then re-immersed at 1430° C., but this time the reading fluctuated wildly.

A further pair of probes was made and dipped at 1305° C. for 15 sec., a steady reading being obtained. The wires were then taken out of the tubes, the short iron-coated lengths at the end were discarded, and a normal welded junction was made. When subjected to a standard salt-point test (sodium sulphate, m.p. 883·5° C.), the thermocouple was shown to be reading 11° C. low, presumably owing to contamination.

Further experiments showed that bare platinum wire dissolved in 13 sec. in molten iron at 1410° C., and almost instantaneously at 1550° C.

It was then decided to carry out a few further trials, but this time the wires were embedded in a refractory block and a minimum surface area of rare metal was exposed to the molten bath—conditions similar to those when the wires are embedded in a furnace lining. Sillimanite powder, bonded with ethyl silicate, was used to form refractory bricks with a diameter of $1\frac{3}{4}$ in. and a length of 10 in. The wires were embedded along the length of the bricks 1 in. apart, and protruded slightly beyond the face. A series of tests was made in the 1500–1600° C. temperature range, immersing the end of the brick $1\frac{1}{2}$–2 in. into the slag-free metal. In no case was a steady reading obtained. On each occasion the protruding wires were dissolved, and the molten iron tended to erode the refractory and to form a small cavity in the block where the wire had been. These holes then slagged over, thus breaking the circuit.

A final test was made, using, for convenience, an alumina cement as the refractory; in other respects conditions were the same as in the previous tests. The bath temperature was 1608° C., and the block was immersed to a depth of 2 in. and held for 1 min. No reading was obtained. Examination of the probe after cooling showed that the platinum wire had been dissolved back $\frac{1}{32}$ in. beyond the face of the refractory, and the position of the platinum–13% rhodium wire was indicated by a small metal/slag head. On stripping the thermocouple it was seen that the platinum wire also had a metal/slag head at its extremity. The platinum wire was rather brittle, and a short length, 0·25 in., broke off. The wires were then examined spectrographically at 0·3-in. intervals, commencing at the immersed end.

Figure 1 shows the estimated amounts of the major contaminating elements present along the wire, based upon the results of the examination. An outstanding point is that the contaminants have diffused back at least $1\frac{1}{4}$ in. along the wire during the immersion time of 60 sec. As this portion coincides with the zone of steepest temperature gradient, the effect upon readings would be great. After the spectrographic examination, a welded junction was made approximately $1\frac{1}{4}$ in. back from the original ends of the wires, and the couple was submitted to a standard salt-point determination (sodium sulphate, m.p. 883·5° C.). It was shown to be reading 11° C. high.

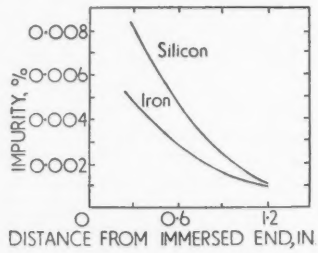

Fig. 1—Contamination of platinum wire, estimated from spectrographic examination

CONCLUSIONS

It is apparent that the use of platinum/platinum–13% rhodium thermocouples without suitable protection from the molten metal is not practicable, as the rate of solution is too rapid and the contamination is too great to allow even rapid readings to be taken.

Acknowledgment

The author thanks the Sheffield Smelting Co. for assistance with the spectrographic examination.

Manuscript received 11th October, 1949.
Mr. Searby, formerly with the Steel Castings Division of B.I.S.R.A., is now at Millspaugh, Ltd., Sheffield.

The Overheating and Burning of Steel

Part III—THE INFLUENCE OF EXCESSIVE REHEATING TEMPERATURES ON THE MECHANICAL PROPERTIES AND THE STRUCTURE OF ALLOY STEELS

By A. Preece, M.Sc., F.I.M., J. Nutting, B.Sc., Ph.D., and A. Hartley, B.Sc.

SYNOPSIS

Mechanical tests have been used to follow the changes induced in a range of alloy steels after treatment at temperatures up to 1425° C. A survey has been made of the influence of cooling rate on the development of the overheated condition. A statistical examination of the effect of steel composition upon overheating temperature has indicated that carbon and sulphur are important constituents, but the influence of oxygen is doubtful. A number of related problems, such as austenite grain growth within the range of overheating temperatures and the influence of overheating on the isothermal decomposition of austenite, have also been investigated.

The conclusions drawn are that overheating is a result of some change in composition at the boundaries of the austenite grains existing at the high temperature. Burning occurs as a result of partial melting and a subsequent segregation of phosphorus, accompanied by the precipitation of sulphides from the molten metal.

Introduction

THE investigations described in the present paper are an extension of some earlier work[1] reported at the symposium[2] on overheating and burning of steel, in 1946.

Some of the earlier experiments have been repeated in more detail and over a wider temperature range so as to provide further information on certain controversial problems which were raised at the symposium. Difficulties were found, for example, in differentiating between the overheated and the burnt conditions, and no generally accepted method was available for distinguishing between the two. Additional information has now been obtained on this point by extending the earlier work to temperatures up to 1425° C. so as to include both overheating and burning.

Certain differences in the results obtained by various workers were thought to be due to the wide variety of cooling rates to which the steels were subjected. It seemed advisable, therefore, to carry out a comprehensive examination of the influence of this factor on the development of the overheated condition with cooling rates ranging from 3 to 3000° C./min.

Earlier work indicated that electric steels had lower overheating temperatures than had open-hearth steels. This suggested that steel composition, particularly with regard to those constituents affected by the steelmaking technique, must exert an important influence on the development of overheating. A complete statistical examination has therefore been made of the effect of composition upon overheating temperature, in an effort to identify the important elements.

As subsidiary investigations, autenite grain growth within the overheating range, and the influence of high-temperature heating on the isothermal precipitation of ferrite from austenite, have been studied. This work, although not complete, shows interesting phenomena not connected with overheating.

An examination of the number and distribution of non-metallic inclusions before and after overheating has been undertaken in view of the observed positive correlation, noted in the earlier paper,[1] between overheating temperature and inclusion count.

Although many factors influencing the development of the overheated condition have been examined, it is still impossible to give a satisfactory explanation of the process, which will account for all the observed phenomena.

INFLUENCE OF HIGH REHEATING TEMPERATURES ON THE MECHANICAL PROPERTIES OF HEAT-TREATED STEELS

The authors' earlier investigations[1] covered temperatures up to 1375° C.; this limit has now been extended to 1425° C. so as to exceed the overheating range and to produce burnt structures. As in the previous work, 1-in. dia. bars were heated to the initial high temperature in approximately 30 min., were soaked for 30 min., and were allowed to cool in air. The bars were then oil-quenched and tempered before testing. All specimens were oil-quenched after tempering.

At each stage in the above series of treatments small specimens were removed from the bar for fracture tests and micrographic examination. After the final treatment (tempering), the remaining length of the bar was used for an izod or tensile test.

All the specimens were heated in a muffle furnace in a stream of nitrogen containing 2% O_2 and 2% H_2O. This slightly oxidizing atmosphere did not produce

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Mr. Preece is Professor of Metallurgy at King's College, University of Durham, Dr. Nutting is Demonstrator in Metallurgy at the University of Cambridge, and Mr. Hartley is at the University of Leeds.

large amounts of fluid oxide at the high temperatures.

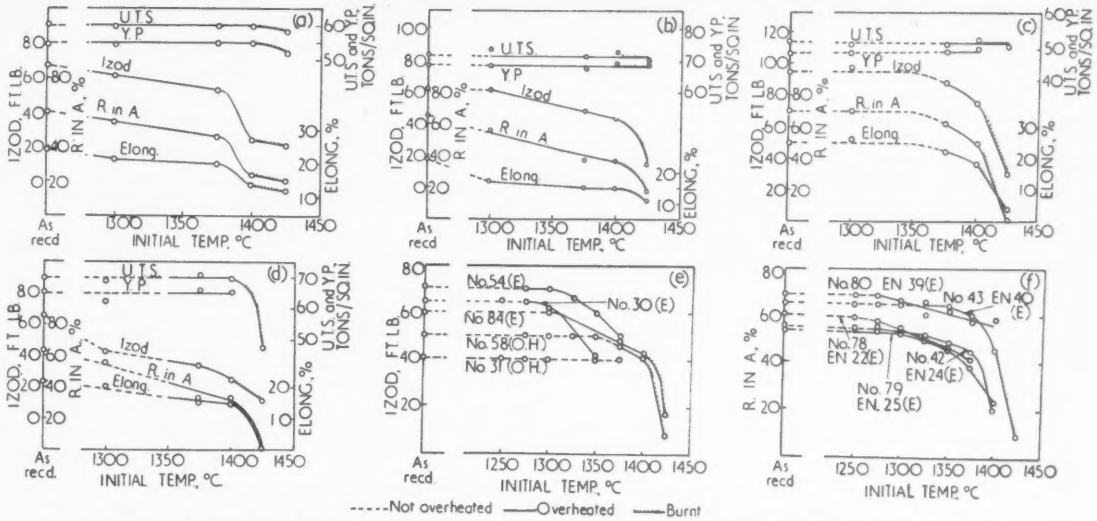
Six different types of steel were used, and in some cases more than one example of each. Both open-hearth and electric steels were included. The compositions and overheating temperatures of the steels used are given in Table I. In Fig. 1 the mechanical properties of the steels in the final heat-treated condition are plotted against the reheating temperature. The curves are drawn so as to indicate the range of temperatures which produce the overheated and the burnt conditions. It is possible to distinguish between the two conditions by the reversal in the

Table I

COMPOSITION AND OVERHEATING TEMPERATURE OF STEELS USED IN THE PRESENT INVESTIGATIONS

Sample No.	En No.	Analysis, %										Method of Manufacture	Overheating Temp., °C.
		C	Si	Mn	S	P	Ni	Cr	Mo	O_2	N_2		
24	40	0·21	0·31	0·56	0·014	0·012	0·20	3·01	0·52	0·004	0·010	E	1300
25	40	0·19	0·27	0·53	0·014	0·013	0·31	2·96	0·41	...	...	E	1300
30	25	0·30	0·30	0·60	0·011	0·020	2·62	0·82	0·57	0·002	0·006	E	1275
31	25	0·30	0·13	0·58	0·030	0·033	2·82	0·62	0·54	0·013	0·004	OH	1450
32	3	0·31	0·26	0·59	0·029	0·020	...	...	...	...	...	OH	1425
33	16	0·38	0·30	1·55	0·029	0·030	...	...	0·27	...	...	OH	1425
34	24	0·40	0·21	0·56	0·021	0·031	1·46	1·54	0·25	0·008	0·008	E	1300
35	31	1·08	0·29	0·47	0·023	0·036	0·19	1·46	...	0·003	0·006	OH	1300
40	39	0·13	0·28	0·45	0·010	0·017	4·42	1·16	...	0·003	0·011	E	1275
41	22	0·39	0·24	0·64	0·016	0·012	3·68	0·30	...	0·005	0·010	E	1300
42	24	0·38	0·30	0·60	0·014	0·014	1·50	1·20	0·16	0·006	0·007	E	1275
43	40	0·27	0·25	0·59	0·018	0·014	0·26	3·13	0·53	...	...	E	1325
46	45	0·57	1·90	0·91	0·050	0·030	...	...	...	0·002	0·005	AOH	1375
47	47	0·57	0·34	0·66	0·040	0·033	...	1·02	0·18V	0·003	0·006	AOH	1375
49	13	0·21	0·14	1·38	0·026	0·026	0·60	0·12	0·26	...	...	OH	1400
50	39	0·17	0·17	0·46	0·013	0·011	4·11	1·35	0·24	0·004	0·007	E	1300
51	40	0·19	0·26	0·50	0·015	0·010	0·35	3·01	0·40	0·003	0·012	E	1300
52	26	0·43	0·28	0·62	0·010	0·012	2·78	0·65	0·60	0·003	0·008	E	1300
53	26	0·43	0·21	0·72	0·029	0·028	2·43	0·66	0·60	0·005	0·003	AOH	1300
54	25	0·30	0·23	0·59	0·014	0·015	2·63	0·68	0·57	0·003	0·007	E	1275
55	39	0·18	0·25	0·49	0·038	0·031	4·17	1·22	0·26	0·009	0·006	AOH	1375
56	31	1·01	0·21	0·40	0·034	0·038	0·12	1·40	0·08	...	...	OH	1300
57	31	0·98	0·26	0·44	0·015	0·030	0·29	1·44	0·04	...	...	E	1300
58	25	0·32	0·20	0·64	0·037	0·033	2·63	0·59	0·50	0·009	0·006	OH	1375
59	40	0·25	0·19	0·47	0·020	0·013	0·25	3·07	0·53	0·006	0·008	E	1350
60	33	0·16	0·21	0·47	0·035	0·018	3·24	0·25	...	0·008	0·004	OH	1375
62	24	0·36	0·34	0·52	0·026	0·011	1·42	0·96	0·27	...	...	HF	1350
63	22	0·39	0·18	0·67	0·036	0·030	3·40	0·21	...	0·004	0·008	AOH	1325
64	33	0·15	0·28	0·50	0·010	0·011	3·00	0·18	...	0·003	0·009	E	1325
65	24	0·37	0·23	0·59	0·023	0·031	1·56	1·15	0·28	0·004	0·008	E	1350
66	39	0·16	0·13	0·45	0·015	0·011	4·31	1·29	0·28	0·004	0·008	E	1325
67	39	0·16	0·13	0·45	0·015	0·011	4·31	1·29	0·28	...	...	E	1325
68	39	0·17	0·24	0·33	0·015	0·013	4·22	1·25	0·22	0·002	0·008	E	1275
69	39	0·17	0·24	0·33	0·015	0·013	4·22	1·25	0·22	...	...	E	1275
70	2A	0·11	0·121	0·45	0·034	0·019	...	...	...	...	...	OH	1400
71	1A	0·13	Trace	1·02	0·226	0·056	...	...	...	...	...	OH	1400*
72	...	0·14	Trace	1·10	0·212	0·046	0·19Pb	...	...	...	...	OH	1400*
73	...	0·35	0·207	1·53	0·043	0·052	0·17Pb		0·22	...	...	OH	1400
74	...	0·25	0·112	1·22	0·027	0·020	0·16Pb			...	...	OH	1400
76	24	0·48	0·23	0·81	0·012	0·032	1·92	0·85	0·33	0·003	0·010	E	1300
77	3	0·18	0·26	0·68	0·027	0·032	0·28	0·18	...	0·003	0·009	E	1400
78	22	0·36	0·13	0·60	0·014	0·011	3·54	0·19	...	0·003	0·010	E	1275
79	25	0·31	0·17	0·64	0·014	0·016	2·53	0·61	0·58	0·003	0·010	E	1250
80	39	0·18	0·19	0·44	0·019	0·020	4·15	1·16	0·27	0·004	0·007	E	1300
81	3	0·13	0·27	0·56	0·009	0·015	0·24	0·16	...	...	...	E	1350
83	22	0·37	0·22	0·62	0·020	0·010	3·32	0·23	...	...	...	E	1275
84	25	0·31	0·24	0·69	0·015	0·012	2·47	0·64	0·56	...	...	E	1300
85	39	0·12	0·21	0·43	0·020	0·015	4·29	1·11	0·20	...	...	E	1375
89	40	0·32	0·32	0·48	0·012	0·029	0·27	3·18	0·57	...	...	E	1350

*Steel burnt at 1400° C. No signs of overheating at lower temperatures.



(*a*) Steel 83, En 22 (electric), 3% Ni–Cr
(*b*) Steel 84, En 25 (electric), 2½% Ni–Cr–Mo, medium-carbon
(*c*) Steel 85, En 39 (electric), 4¼% Ni–Cr, case-hardening
(*d*) Steel 89, En 40 (electric), 3% Cr–Mo, nitriding
(*e*) En 25 (electric and open-hearth), 2½% Ni–Cr–Mo, medium-carbon
(*f*) A range of alloy steels

Fig. 1—Influence of high reheating temperature upon mechanical properties

etching properties of the steels. This is dealt with later in the paper.

In considering the results shown in Fig. 1, it will be noted that whilst there are differences in the behaviour of each steel the following generalizations are possible :

(*a*) The tensile strength and yield point of the medium- and low-carbon alloy steels in the finally heat-treated condition (up to 0·4 % C) are not appreciably affected by prior exposure to reheating temperatures as high as 1425° C.

(*b*) In the same way ductility and toughness, as indicated by reduction in area and the izod value respectively, do not fall appreciably until reheating temperatures of 1375° C. have been attained. With temperatures in the region of 1400° C. and above, serious deterioration in these mechanical properties may occur.

It has been shown previously that the overheated condition can be detected by fracture and etch tests. The evidence revealed by the former becomes more pronounced the greater is the degree of overheating. The etch test, on the other hand, gives only a positive or a negative indication. It must be emphasized that neither of these tests gives a reliable indication of the degree to which the steel has become embrittled due to overheating. For example, with steels 83 and 84, Figs. 1*a* and *b*, the transverse fracture test after treatment at 1350–1375° C. was completely granular and had the appearance of a brittle metal, yet, as will be seen from Figs. 1*a* and *b*, the mechanical properties were not seriously impaired.

This difficulty in assessing the deterioration in properties by a simple fracture test has been partially solved by Haworth and Christian,[3] who claim as a result of their work that the mechanical properties of overheated steel are not seriously reduced until facets become visible in a longitudinal fracture test. A further useful guide which the present authors have found in their examination is that, as long as the transverse fracture test shows the presence of a few areas of a fibrous character amongst the facets, the properties of the steel are not appreciably affected.

Microstructural Changes

Whilst the etch test does not reveal the intensity of overheating, it can be used for detecting the presence of the overheated structures. Its most useful application, however, especially in view of the fall-off in the curves in Fig. 1, is in the detection of the onset of burning.

Up to the present, the only evidence available for deciding whether or not a steel was burnt was the presence of certain visible defects in a polished specimen, such as voids, intercrystalline separation, or the presence of oxide envelopes around the grain boundaries. These are obviously unsatisfactory criteria since the structural disturbances must reach certain proportions before they become visible. A more satisfactory indication of the burnt condition can be obtained from the change in etching characteristics which accompanies the first stages of burning. It has been found that the action of both the electrolytic ammonium nitrate and the nitro-sulphuric etchants changes in a spectacular manner as the steel becomes burnt. These changes in the appearance of the etched surface occur before the structural defects, such as intercrystalline separation and oxide envelopes, become visible.

The changes which occur in the action of the etchants in the transition from the overheated to the burnt structure are as follows :

10% *Nitric*–10% *Sulphuric Acid*—In the overheated condition the boundaries of the pre-existing austenite grains are attacked and therefore appear black (Fig. 2). With the onset of burning, the grain is attacked in preference to the boundaries and the microstructure shows a system of white boundaries with a dark background (Fig. 3).

Electrolytic Ammonium Nitrate—The results obtained are directly opposite to those obtained with the nitro-sulphuric reagent, *i.e.*, a white boundary network is seen in the overheated condition (Fig. 4) and a black network in the burnt state (Fig. 5).

2% *Nital and Saturated Picral*—With overheated steels these etchants, singly or together, give the normal sorbitic structure, but with burnt steels nital attacks the grain in preference to the boundary, producing white boundaries on a dark background (Fig. 6), while picral gives the opposite effect (Fig. 7).

If these two reagents are mixed their action depends upon the relative concentrations and the etching time. This fact accounts for the confusion existing in many published photomicrographs of burnt steels, where in some cases black boundaries are shown and in others they are white.

The reversal in etching properties must be a result of a change in composition of material at the austenite grain boundary, presumably due to incipient fusion. The liquid phase formed would be enriched in carbon, sulphur, and phosphorus, and in cooling the carbon would diffuse back into the austenite so that the remaining segregation would be chiefly that of sulphur and phosphorus. Steel 89 (0·4% C, 3% Cr), heated to 1425° C., showed clear evidence of dendrites in the central zone of the fractured surface, indicating that fusion had occurred in the centre of the bar. This dendritic structure is shown in Fig. 8. Figure 9 shows the structure after polishing and etching in electrolytic ammonium nitrate, when the interdendritic area etches black. On moving from the centre, which had been molten, to the outer portion of the bar, which had been only partially molten, the black interdendritic areas become less and merge into the black boundaries of burnt steel. This establishes the fact that the black boundaries shown by the burnt steel (ammonium nitrate etchant) are the result of incipient fusion.

This reversal in etching properties clearly indicates the change from the overheated to the burnt condition and offers a satisfactory test for distinguishing between the two. It also serves as a basis for a simple test in detecting burnt steel, since a reasonably polished surface will show the coarse black boundary structure, with the electrolytic ammonium nitrate etch, sufficiently clearly for it to be seen with a good pocket lens.

It is noted that, with the steels listed in Figs. 1*a* to *d*, the reversal in etching properties occurred after an initial heating to a temperature between 1400° and 1425° C., whilst the rapid fall in mechanical properties commenced after heating to temperatures between 1375° and 1400° C. Had the temperature intervals in the experiments been less than 25° C., there may have been closer agreement between the temperature necessary to give a reversal in etching properties and that which causes rapid deterioration in mechanical properties. Even so, it is evident that the rapid fall-off in properties precedes, by a small margin, the change in etching.

Thus, although the reversal in etching indicates a burnt condition and poor mechanical properties, it cannot be assumed that no deterioration has occurred if the etched structures show only the overheated condition. This is unfortunate, since, if the fall-off in mechanical properties coincided exactly with the reversal in etching properties the latter would then constitute a most useful non-destructive test in the examination of forgings, etc.

INFLUENCE OF COOLING RATE ON THE DEVELOPMENT OF THE OVERHEATED CONDITION

It has been shown that intermediate rates of cooling are more favourable for the production of the overheated condition than are either slow or rapid cooling rates. This suggests that the minimum temperature necessary to overheat a steel depends on the subsequent rate of cooling, and it seemed worth while to examine in more detail this relationship between the cooling rate and the overheating properties of the steel.

For this examination six cooling rates were adopted:

Cooling Rate, ° C./min.	Method of Cooling
3	Furnace-cooled
10	,,
20	,,
425	Air-cooled
1300	Oil-cooled
3000	Water-cooled

The above values are the average cooling rates* in the temperature range 1375–1200° C., and were measured by a thermocouple in the centre of a bar of the same dimensions as the specimens used for this work, *i.e.*, 1-in. dia. and 6 in. long.

In view of the large number of experiments involved, this detailed investigation was limited to three steels:

Steel	Overheating Temp.,† ° C.
83	1275
84	1300
85	1375

† Air-cooled (*i.e.*, 425° C./min.) from initial high temperature.

The experimental procedure was similar to that previously employed. Specimens of the three steels were heated to temperatures which varied by steps of 25° C. Samples were cooled from each temperature at each of the six cooling rates. After the initial treatment all the specimens were given the appropriate oil-quenching and tempering treatment. Fracture tests were taken (i) after cooling from the initial high temperature, (ii) after oil-quenching, and (iii) after tempering. Etch tests were carried out on polished surfaces of samples from the fracture specimens and by this means a comparison was made of the sensitivities of the fracture test and the etch tests.

The results obtained are shown in Fig. 17. The curves indicate the minimum temperature at each cooling rate which produced evidence of overheating. For convenience, this temperature is referred to as the overheating temperature. With a slow cooling rate, higher temperatures could be tolerated before producing the overheated condition. This is shown

* The cooling rates in oil and water were determined by Mr. R. D. Carter. No high degree of accuracy is claimed for these measurements, but they are sufficiently accurate for the present requirements.

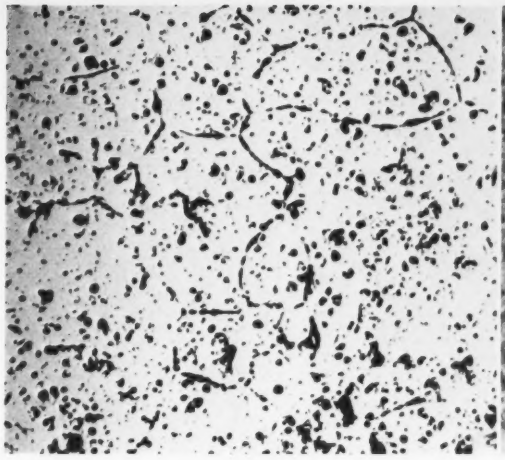


Fig. 2—Overheated steel etched with 10% nitro-sulphuric acid, showing black boundaries × 50

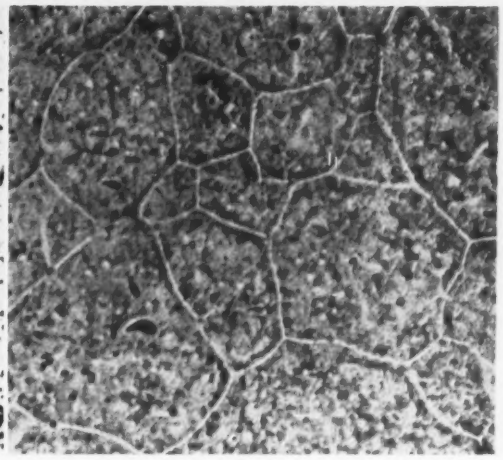


Fig. 3—Burnt steel etched with 10% nitro-sulphuric acid, showing white boundaries × 50

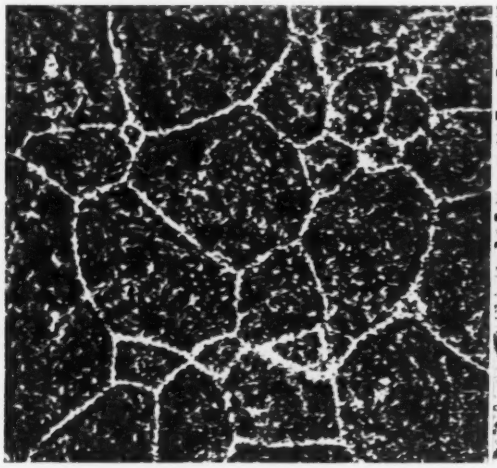


Fig. 4—Overheated steel etched electrolytically with saturated ammonium nitrate solution, showing white boundaries × 50

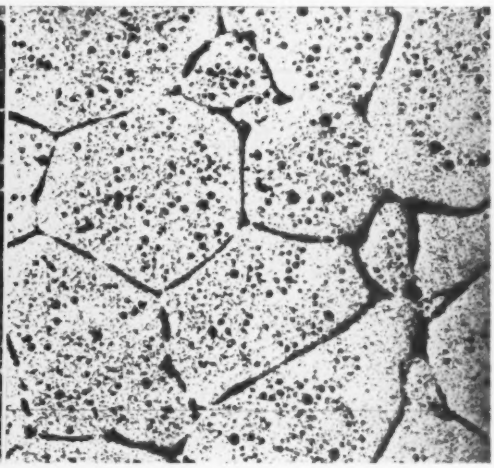


Fig. 5—Burnt steel etched electrolytically with saturated ammonium nitrate solution, showing black boundaries × 50

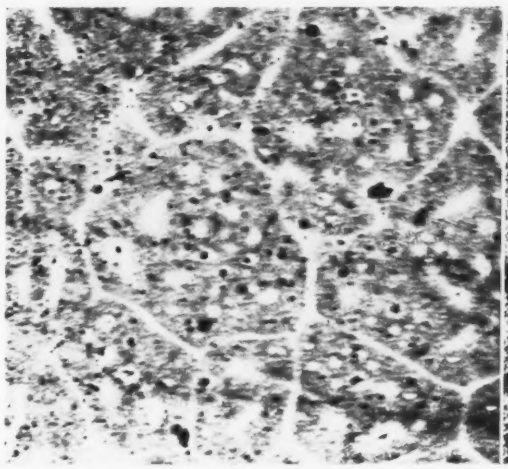


Fig. 6—Burnt steel etched with 2% nital, showing white boundaries × 50

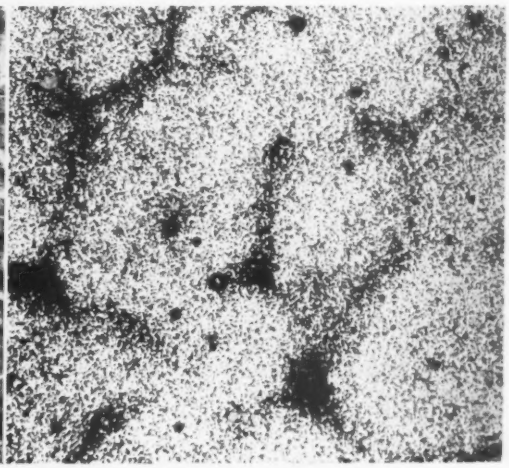


Fig. 7—Burnt steel etched with saturated picral, showing black boundaries × 50

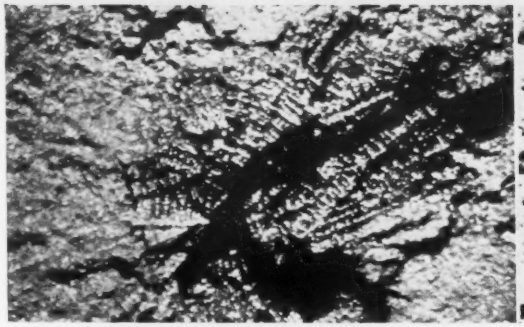


Fig. 8—Fracture of steel 89 after treatment at 1425° C., showing dendrites × 20

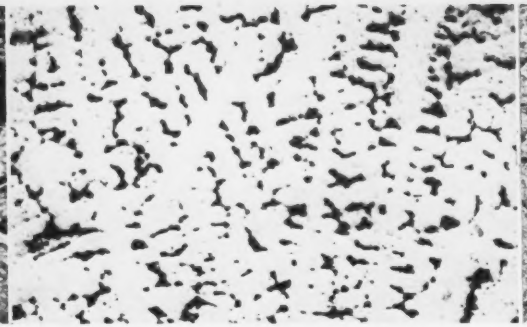


Fig. 9—Microstructure of steel 89 after treatment at 1425° C. Etched electrolytically with saturated ammonium nitrate solution × 75

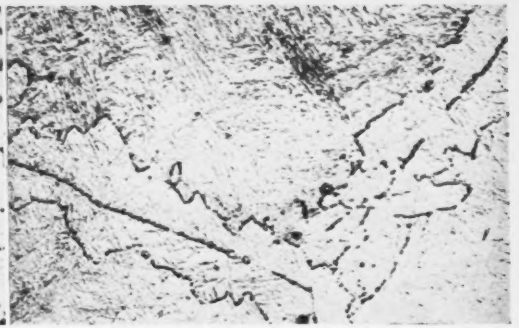


Fig. 10—Microstructure of steel 80 after treatment shown in Fig. 19a. Etched with 2% nital × 150

Fig. 11—Microstructure of steel 80 after treatment shown in Fig. 19b. Etched with 2% nital × 150

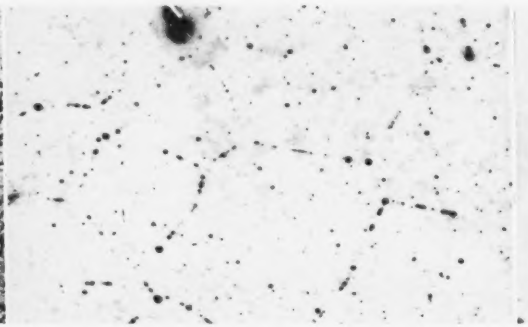


Fig. 12—Photograph of sulphur print from a burnt steel × 25

Fig. 13—Fine dispersion of sulphides obtained in free-cutting steel after water-quenching from 1425° C. × 1500

Fig. 14—Large globular sulphides obtained in free-cutting steel after treatment at 1425° C., followed by cooling at 3° C./min. to 1300° C. × 1500

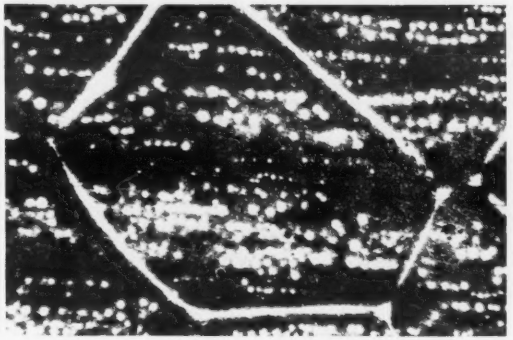


Fig. 15—Burnt steel etched in Oberhoffer's reagent, showing phosphorus segregation × 75

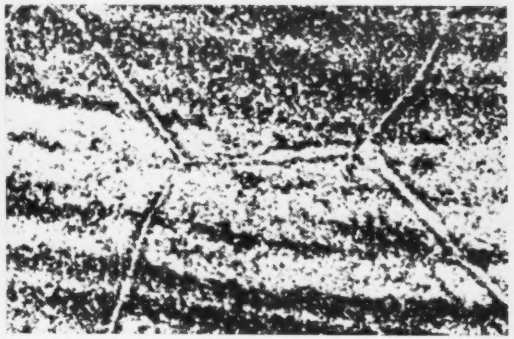


Fig. 16—Overheated steel showing flow lines and white boundaries. Etched electrolytically with saturated ammonium nitrate solution × 75

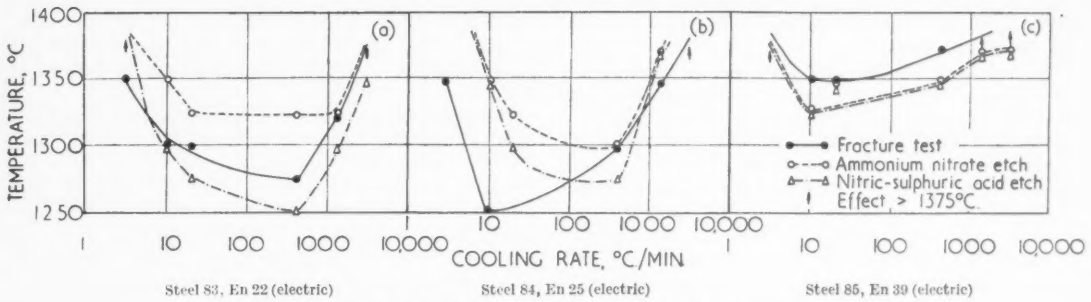


Fig. 17—Effect of cooling rate upon overheating temperature

by the rise in the curves. Rapid rates of cooling had a similar tendency, but were not as effective as slow cooling in suppressing overheating.

The greatest susceptibility to overheating was associated with cooling rates between 10 and 400° C./min. These are rates of cooling usually found in industry, and the shallowness of the trough shown by the curves indicates that, as far as overheating is concerned, mass effect is of no importance unless the sections are extremely large.

The curves in Fig. 17 give the results obtained with the steels after full heat-treatment, *i.e.*, oil-quenched and tempered. Similar series of curves were obtained from tests carried out on specimens in the other stages of the heat-treatment, *i.e.*, after cooling from the initial high temperature and in the fully hardened condition. Consistent results were obtained only after the final tempering operation. With fast rates of cooling, much confusion must inevitably arise if the steel is not fully heat-treated before testing. This applies to both the fracture and the etch tests. For example, granular fractures were obtained after oil- and water-quenching from quite low temperatures (1100° C.), due presumably to quenching stresses, but certainly not due to overheating. Etch tests were similarly misleading. Tests carried out in the fully hardened condition were also unsatisfactory.

In general, the curves in Fig. 17 show that the fracture test is the most reliable method for detecting the early stages of overheating. The nitro-sulphuric acid etch, although sensitive, is not entirely satisfactory since the clarity of the structure obtained depends upon the degree of repolish after etching. The ammonium nitrate etch, although giving a clear indication of overheating, is not as sensitive as the fracture test.

GRAIN GROWTH AT TEMPERATURES WITHIN THE OVERHEATING RANGE

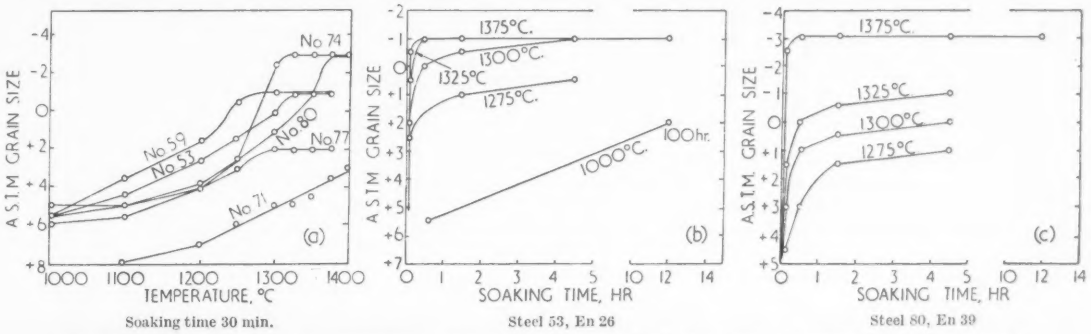
In examining the fractured surfaces of overheated steels it was noticed that specimens which had been heated for prolonged periods did not indicate appreciable grain growth, a fact which seemed unusual. An examination of the literature gave no information on grain growth up to temperatures of 1300° or 1400° C.; recent investigations by Whiteley[4] did not go beyond 1200° C.

The influence of temperature and of soaking time was examined, and the results are shown in Fig. 18.

For this work, $\frac{1}{2}$-in. dia. × $\frac{3}{4}$-in. specimens were heated to temperature in approximately 5 min., held for 30 min., and quenched in water. Vilella's etching reagent was used for revealing the austenite grains, and the grain size was estimated by comparison with standard A.S.T.M. charts.

A steady increase in grain size up to approximately 1150° C. was obtained with a wide variety of steels. Above this temperature, growth became very rapid and a grain size was quickly attained, which, with a number of steels, remained almost constant with further increase in temperature. This is illustrated in Fig. 18*a*.

Steel 71 (a free-cutting steel and possessing a smaller grain size than the other steels) behaved rather differently from the others. At 1300° C. dominant growth began in a few grains whilst the rest of the grains remained at approximately constant size with increasing temperature. The curve in Fig. 18*a* for this steel shows the increase in the dominant grains



Soaking time 30 min. Steel 53, En 26 Steel 80, En 39

Fig. 18—Influence of (*a*) temperature and (*b*) and (*c*) soaking time upon austenitic grain growth

only. With the other steels the grain size was fairly uniform after heating.

The results were contrary to expectation and did not give the large grain size occasionally found in practice. The effect of soaking time was then investigated, with the results shown in Figs. 18*b* and *c*. A rapid rate of grain growth, which increased with temperature, occurred, enabling the steel to attain very quickly a grain size which, in the region of 1375° C., was not appreciably affected by soaking for more than 30 min. At lower temperatures this initial period of rapid grain growth was still present, but to a lesser degree, so that further soaking had a more appreciable influence on grain size.

It will be appreciated that in estimating the grain size in these experiments, an average value has been taken of a number of fields. The actual values obtained did not lie on smooth curves, as shown in Fig. 18, but the scatter was slight so that, within the experimental errors inherent in the method, the curves indicate fairly accurately the average grain size obtained with each steel.

There is apparently no connection between the overheating temperature and the temperature at which cessation of grain growth occurs.

Since the maximum grain size attained in the experiments was much less than that occasionally found in industry, further specimens were subjected to the following treatments, in an attempt to produce larger crystals :

(*a*) Heated to 1375° C. at various slow rates and soaked for varying times with intermediate soaking periods at 950°, 1150°, and 1300° C.

(*b*) Heated to 1375° C. to develop 'maximum' grain size, cooled (water-quenched, etc.) and again soaked at 1375° C.

(*c*) Heated to 1375° C., as in (*b*), then hot-worked in the austenitic zone and again soaked at 1375° C.

None of these treatments produced a grain size larger than the 'maximum' developed by a brief soak at 1375° C.

In a further series of experiments the influence of cyclical heating and slow cooling in the austenitic zone was investigated. A treatment at 1375° C. followed by slow cooling to 1250° C. and then reheating to 1375° C. was carried out three times, and by this means the grain size of steel 77 (plain carbon, electric) was increased from + 2 to — 4 A.S.T.M. This suggests that a large grain size may be produced industrially by temperature fluctuations.

The results are of interest in considering the mechanism of overheating and they explain why the faceted structure of an overheated steel does not become coarser on prolonged soaking.

EFFECT OF HIGH TEMPERATURES UPON THE SUBSEQUENT AUSTENITE–FERRITE REACTION

Since overheating results from a change in composition of the austenite grain boundary relative to the main body of the grain, it was thought that this change might influence the precipitation of ferrite from austenite.

Specimens were heated above and below their respective overheating temperatures and the isothermal decomposition of the austenite to ferrite was examined after the initial heating and also after imposing an oil-quenching and tempering treatment between the initial heating and the isothermal transformation. The treatments are shown in Fig. 19.

The procedure adopted for examining the isothermal change was the same as that used by Bain,[5] *i.e.*, the specimens were heated at 870° C. for 10 min., plunged into a lead bath at 650° C., held for the required period, and then quenched in water. The products of the decomposition were ferrite, produced at 650° C., and martensite formed by quenching the undecomposed austenite. The temperature of 650° C. was chosen since it was found to give the maximum precipitation of ferrite. Reaction times varied up to 100 hr.

With the treatment outlined in Fig. 19*a*, ferrite precipitation occurred along the austenite grains established at the initial heating A and at all values of A, whether below or above the overheating temperature. The microstructure of a typical specimen is shown in Fig. 10. No differences were observed in microstructures as the temperature of the initial heating passed through the overheating temperature.

When an oil-quenching and tempering treatment was interposed between the initial heating and the isothermal transformation, *i.e.*, A' B' C' D' E', Fig. 19*b*, different results were obtained. Ferrite precipitation occurred at the austenite boundaries established at B' and also at the boundaries existing at A′, but only after the initial heating temperature had exceeded a certain value, which varied between 1250 and 1300° C. for the steels used in this study. Figure 11 shows the ferrite outlining the two sets of austenite grains. No significance is attached to the temperatures at A', which produce austenite boundaries sufficiently persistent to influence ferrite precipitation in the isothermal transformation, since these temperatures were well below the overheating temperatures of the respective steels and were in no way related to them, as is shown by the following :

Steel	Type	Manufacture	Temp. to Initiate Ferrite Precip., ° C.	Overheating Temp., ° C.
80	En 39	Basic electric	1250	1300
55	En 39	Open-hearth	1250	1375
58	En 25	Open-hearth	1300	1375

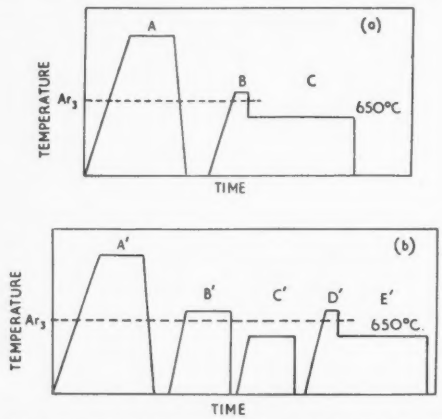


Fig. 19—Thermal history of the specimens used for the study of isothermal precipitation of ferrite

The cooling rate from A' had no influence on the subsequent isothermal decomposition.

It will be evident that the isothermal transformation of austenite to ferrite is not influenced by overheating, but the results obtained in these experiments are of interest in relation to some results published by Preece and Irvine,[6] which showed that carbon diffuses away from areas high in sulphur. The precipitation of ferrite along austenite boundaries established at high temperature may be due to a low carbon content in these regions as a result of sulphide precipitation during cooling.

STATISTICAL EXAMINATION OF STEEL COMPOSITION AND OVERHEATING TEMPERATURE

Since the susceptibility to overheating appeared to be dependent on the composition of the steel and possibly on the steelmaking technique, a statistical examination has been made in an attempt to determine the influence of these factors on the overheating temperature.

The first step was to determine the simple correlation coefficients between overheating temperature and the percentage of each of the elements, using 29 steels for which complete analyses were available. The results are given below. Sulphur is predominant in these results and the conditions of steelmaking normally give rise to a relationship between sulphur and the other elements ; this may be expected to be particularly marked between sulphur and phosphorus. Partial correlation coefficients have accordingly also been determined between overheating temperature and each of the other elements with sulphur constant, and these are included in the following data :

Simple and Partial Correlation Coefficients

Overheating Temperature and :	Simple Correlation Coefficient	Partial Correlation Coefficient, Sulphur Constant
Sulphur	0·7193	....
Oxygen	0·5888	0·4957
Carbon	−0·0269	−0·3221
Silicon	0·1954	−0·2467
Manganese	0·1887	−0·1536
Phosphorus	0·5505	0·1153
Nitrogen	−0·4718	−0·0549

Of the partial correlation coefficients, that for oxygen is highly significant ; that for carbon is not significant at the 5% level, but it may be considered doubtful. The remainder are not significant even at levels well over 10%. Subsequent statistical analysis is therefore confined to the elements sulphur, oxygen, and carbon.

To determine whether the method of steelmaking should be included as an important variable, the steels have been divided into two groups, electric and open-hearth. Unfortunately oxygen analyses were available for only nine of the open-hearth steels. Further statistical analysis has therefore been made in two parts :

(*a*) Effects of sulphur, oxygen, and carbon in the 20 electric steels and nine open-hearth steels for which oxygen analyses were known. (These are the 29 steels from which the previous coefficients were calculated.)

(*b*) Effects of sulphur and carbon in 46 steels (30 electric and 16 open-hearth steels).

The partial correlation and partial regression coefficients for sulphur and oxygen and for sulphur and carbon for the 29 steels with oxygen analyses, are as follows :

Partial Correlation and Partial Regression Coefficients

Partial Correlation Coefficient	Electric Steels $N = 20$	Open-Hearth Steels $N = 9$	Electric + Open-Hearth Steels $N = 29$
$r_{TO \cdot S}$	−0·169	+0·840	+0·496
$r_{TC \cdot S}$	−0·262	−0·463	−0·322
Partial Regression Coefficient			
$b_{TO \cdot S}$	−3112	+10,343	+2846
$b_{TS \cdot O}$	+5639	+3303	+2846
$b_{TC \cdot S}$	−59·3	−77·3	−57·2
$b_{TS \cdot C}$	+5233	+1163	+3386

It will be observed that the partial regression of temperature on oxygen is positive and that the partial correlation coefficient is significant for the open-hearth group and for the total combined ; in the electric group the regression is negative and the partial correlation coefficient is non-insignificant. There is clearly a contradiction here. It is considered, however, that the number of open-hearth steels is too few for it to be said with certainty that there is a real difference between the two types of steel, and in view of the complete lack of significance with the electric steels, the apparent effect of oxygen on overheating temperature must be considered doubtful.

The effect of carbon is consistent between the two groups of steels, and its influence has been determined on 46 steels taken from Table I. (Analyses of 49 steels are shown in Table I ; steels 71 and 72, which showed signs of burning and no indication of overheating, and steel 62, which was made in a high-frequency furnace and therefore was difficult to classify, were omitted from the statistical analysis.)

The partial correlation coefficients and partial regression coefficients for the 46 steels are given below. The partial correlation coefficient between carbon and overheating temperature is − 0·392 ; this is a highly significant value when 46 observations are considered and it compares with the non-significant value of − 0·322 found earlier when considering the 29 steels with oxygen analyses.

Partial Correlation and Partial Regression Coefficients

Partial Correlation Coefficient	Electric + Open-Hearth Steels $N = 46$
$r_{TC \cdot S}$	−0·392
$r_{TS \cdot C}$	+0·703
Partial Regression Coefficient	
$b_{TC \cdot S}$	−70·9
$b_{TS \cdot C}$	+3491

The following conclusions may be drawn from the results obtained in the statistical analysis :

(1) There is no evidence that elements other than sulphur, oxygen, and carbon affect the overheating temperature of a steel.

(2) Owing to oxygen analyses not being available for a large number of steels, it is not possible to obtain a clear picture of the combined effect of the three elements.

(3) Considering the 29 steels for which oxygen analyses were available, there is a significant partial correlation between oxygen and overheating temperature. The breakdown of the steels into electric and open-hearth groups shows no significant correlation between oxygen and overheating temperature for electric steels, but a very marked correlation for the open-hearth steels. As there were only nine open-hearth steels, the effect of oxygen must be considered doubtful.

(4) Considering the effects of sulphur and carbon in all the 46 steels, it is found that sulphur is positively correlated with overheating temperature and carbon negatively; *i.e.*, increasing the sulphur content of a steel will increase the overheating temperature, whilst increasing the carbon content will decrease the overheating temperature.

INFLUENCE OF OVERHEATING AND BURNING ON THE DISTRIBUTION OF NON-METALLIC INCLUSIONS IN STEEL

In view of the emphasis on sulphur and oxygen in the statistical examination and also the close relationship between overheating and inclusion count, it seemed worth while to investigate the effect of both overheating and burning on the distribution and morphology of the non-metallic inclusions.

Two specimens from each of a number of steels with overheating temperatures ranging from 1250° to 1425° C. were heated at 1375° C. for 30 min. One series was water-quenched and the other was air-cooled to room temperature. A third series was retained in the as-rolled condition. The air-cooled and the as-received specimens were then hardened by water-quenching from above the A_3 point, and longitudinal sections were prepared for microscopic examination. Results of inclusion counts obtained by two observers showed no significant differences before and after these treatments.

When specimens of the low-sulphur steel 81 were quenched from temperatures between 1100° and 1375° C., there was a slight decrease in the intensity of the sulphur print, the decrease being greater the higher the quenching temperature, indicating a slight increase in the solubility of sulphur in this region. This increase was not sufficient to have any appreciable influence on the inclusion count.

Overheated steels did not show any unusual features in the sulphur print, but burnt steels gave clear evidence of sulphide segregation at the boundaries of the austenite grains. Figure 12 is an enlarged photograph of a sulphur print taken from a burnt steel in which oxide films were visible around the crystals. It is evident, therefore, that these oxide films are rich in sulphur.

It was noticed that with the high-sulphur (0·2% S) free-cutting steel 72, there was no apparent change in the elongated shape of the manganese sulphide inclusions after heating at temperatures up to 1400° C. After treatment at 1425° C. the sulphides were globular and were largely at the austenite grain boundaries. Quenching the specimen from 1425° C. produced a fine dispersion of inclusions at the crystal boundaries (Fig. 13). Slow cooling gave massive globules (Fig. 14).

This change in the distribution of the sulphides indicates a marked increase in the solubility of sulphur at temperatures above 1400° C., and the same may be true for the oxides. It is clear, however, that the solubility of sulphur in low-carbon steels is slight at temperatures below 1400° C. This has been confirmed by quenching specimens of very low-sulphur steel from high temperatures. No appreciable difference was noticed in the size of the sulphide particles.

INFLUENCE OF CARBON ON OVERHEATING TEMPERATURE

Whilst it is extremely difficult to isolate the influence of any one element when studying a particular property in commercial steels, the influence of carbon may be examined since it is possible by carburizing to increase the carbon content of a steel without appreciably changing the content of the other elements.

The overheating temperatures of a number of steels, after carburizing for varying periods and homogenizing, have been examined, with the results shown in Fig. 20.

Attempts to decarburize Ni–Cr and Ni–Cr–Mo steels by heating at 1000° C. in a stream of hydrogen containing 15% H_2O, produced intense brittleness. This brittle condition could not be removed by normalizing nor by prolonged heating at 1000° C. in a vacuum. Fracture occurred along cleavage planes. Failure to remove the brittle condition made this method of varying the carbon content unsatisfactory.

CONSIDERATION OF RESULTS

The loss in mechanical strength of high-tensile alloy steels of low-to-medium carbon content, as a result of excessive reheating temperatures, is not severe until these temperatures approach the range which gives burnt steel. With the steels used in the present investigation the toughness and ductility fall off rapidly after heating in the region of 1375–1400° C., and evidence of burning was found after treatment at temperatures slightly above 1400° C. The overheating temperature for most of the steels is appreciably below this zone and it would seem, therefore, that it is not overheating, but the onset of burning, which is responsible for the loss in strength. This conclusion is also suppor-

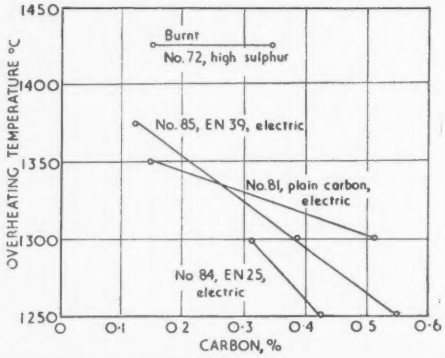


Fig. 20—Effect of increasing carbon content upon overheating temperature (constant cooling rate)

ted by the fact that many examples have been found where the fractured surface of an overheated steel was completely granular, indicating an advanced stage in the overheating, yet the mechanical properties of the metal were unaffected.

Since overheated and burnt steels react differently to certain etching reagents, etch tests may be used to differentiate between the two conditions. It is unfortunate that the transition temperature (from overheated to burnt), as indicated by the change in etching properties, is a little higher than the temperature necessary to bring about the serious fall in mechanical strength. Had the converse been true, the change in the etching properties would have been a most useful non-destructive test for detecting faulty material caused by excessive temperatures. It has a limited application, however, in that it reveals an earlier stage of burning than is shown by such tests as crystal separation, oxide envelopes, etc. With ammonium nitrate the burnt condition is shown as a black grain-boundary network on a light background. The absence of black boundaries, however, cannot be accepted as an indication that the steel is satisfactory, and in such circumstances the fracture test, as outlined in the paper, is the safer guide.

By comparing structures etched with ammonium nitrate with those etched with Oberhoffer's or Stead's reagent, it is evident that the constituent which etches black with ammonium nitrate is the segregated material resulting from incipient fusion. The burnt condition as revealed by Oberhoffer's reagent is shown in Fig. 15.

Further evidence that the constituent which etches black with ammonium nitrate is the segregated material resulting from incipient fusion, was found in etched structures of a specimen which had been heated to such a temperature that the centre of the bar had been molten and, on cooling, formed clear dendrites. The impure material between the dendrites etched black with the ammonium nitrate, and these areas were continuous with the black boundaries in the outer portion of the bar where fusion had only reached an incipient stage.

It will be clear, therefore, that since overheated steels do not give this black-boundary effect on etching with ammonium nitrate, overheating does not involve the melting of the grain boundary and represents a different mechanism from burning.

It was shown in the authors' earlier publication[1] that the overheated condition is developed during the cooling of the steel through the overheating range, *i.e.*, from the high temperature to approximately 1200° C. The results obtained in the present investigation have confirmed this finding and also the fact that the overheating tendencies of a steel can be minimized by very slow or very rapid cooling. In addition, it has been shown that a change in the cooling rate from 10 to 400° C./min. has no significant influence on the minimum overheating temperature, and since this range covers cooling in air for the normal run of forgings, it may be claimed that mass effect, except for very large or very small components, is not important as far as overheating is concerned. However, where cooling rates below 10° C./min. are possible, overheating tendencies are decreased, especially if the cooling rate is reduced to a figure of the order of 3° C./min.

Whilst the information contained in the present paper does not provide an explanation of the mechanism of overheating, it has established certain relevant facts which must be taken into consideration in any attempt to explain this mechanism.

In the first place, it has been found that the solubility of sulphur does not increase appreciably on heating, until temperatures which result in burning are approached ; it is therefore difficult to accept the precipitation of sulphide or oxy-sulphide as the cause of overheating. It has been shown that when temperatures have been sufficiently high to produce the molten phase at the austenite boundary, this phase is clearly revealed by appropriate etching reagents. In the overheated condition, however, these same etchants give quite a different reaction.

It will be appreciated also that the fall in mechanical properties of an overheated steel is not serious, even when the fracture is completely granular in appearance. If this type of fracture were due to the presence of a film of some precipitate, serious reduction in mechanical strength would be expected.

The influence of tempering in developing the faceted or granular fracture is also considered to be of significance since it indicates differences in composition between the material which constitutes the grain boundary and that of the grain itself. A precipitate at the grain boundary would not be expected to influence the tempering properties of the grain boundary. It may be of interest to draw attention to Fig. 16, where the weakened structure is shown to have an appreciable thickness and to be continuous with the rest of the material. It will also be clear from Fig. 16 that the facets in the fractured surface of an overheated steel do not represent crystal boundaries, but merely the position of the crystal boundaries which existed at the reheating temperature.

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The Detection of Overheating and Burning in Steel by Microscopical Methods

By A. Preece, M.Sc., F.I.M., and J. Nutting, B.Sc. Ph.D.

SYNOPSIS

A review is given of the etching reagents which have been used to detect overheating and burning in steel. An investigation of the action of an electrolytic ammonium nitrate etch is described and a theory advanced for its differential action with not overheated, overheated, and burnt steel.

Introduction

OVERHEATING and burning are two defects induced in steel after treatment at forging temperatures. These defects are readily made apparent by fracture tests taken after the steel has been put in its toughest condition, when instead of the normal fibrous appearance the fracture shows evidence of granularity. The incipient stage of overheating is indicated by a faceted fracture. The facets are small zones of granularity which are superimposed upon a background of the normal fibrous structure. As the severity of overheating becomes greater, the ratio of faceted to fibrous zones increases, until the limit is reached when a completely granular fracture is obtained. The fracture of a burnt steel is usually completely granular, but at the same time the facet surfaces appear to have been molten. Burning so weakens a steel that a granular fracture may be obtained in any condition of heat-treatment.

A clearer distinction between overheated and burnt steel may be obtained by microscopical examination, and it is proposed to describe methods which have been developed for this purpose.

OVERHEATING

The detection of overheating by microscopical methods is a problem presenting many difficulties. During overheating, no structural changes which occur are visible under the microscope when the specimen to be examined is in the polished but unetched condition. Neither is it possible to differentiate between overheated and not overheated steel after etching with 2½% nital or saturated picral.

Merchant[1] and Goodrich[2] have claimed that nital can be used to detect overheated steel. But from their published photomicrographs this etchant only shows grain contrast, which is not indicative of overheating since the same effect may be obtained in steels heated to high temperatures, but not overheated.

Merchant also described the use of a macroscopic method involving the pickling of forgings in 50% hydrochloric acid. It seems that this procedure will give only an indication of the austenite grain size established during the high-temperature treatment. The acid attacks the metal preferentially where intercrystalline penetration of the oxide scale has occurred during the soaking treatment.

Prior to the authors' earlier investigations[3] the only satisfactory etchant for the detection of overheated steel was that developed by Slagg and De Belin and reported by Wesley Austin.[4] This etchant consisted of a 10% aqueous solution of nitric and sulphuric acids. The specimen to be etched was immersed in the acid for 30 sec., removed, washed, and the black surface film swabbed away. This process was repeated three times, after which the specimen was lightly polished. The action of the etchant is to attack preferentially the boundaries of the pre-existing austenite grains, giving a system of black grain boundaries with a lightly attacked background (Fig. 1). This etchant differentiates between overheated and not overheated steel, and its action is almost independent of the condition of heat-treatment of the steel.

The nitro-sulphuric acid reagent is unsatisfactory when investigating the causes of overheating, since it dissolves away the metal at the region which is of greatest interest. It was for this reason that Preece[3] and co-workers investigated the possibility of using other reagents for detecting overheating. In all, over 300 reagents were tried, the most satisfactory of which were ammonium persulphate, molten zinc chloride, and an electrolytic etch using a saturated aqueous solution of ammonium nitrate as electrolyte.

The microstructure obtained with each of these reagents, when applied to overheated steel, was a system of white boundaries with a black background (Fig. 2). While the ammonium persulphate etch was convenient in use, it was not as sensitive as the ammonium nitrate etch in detecting the incipient stages of overheating. The molten zinc chloride was sensitive in detecting incipient overheating, but was troublesome in use and required many hours for etching. Attention was confined, therefore, to the ammonium nitrate etchant. With the etching conditions reported earlier the ammonium nitrate was not an entirely satisfactory reagent, since it was not generally applicable. An indication of overheating could be obtained only when the steel was in the fully heat-treated condition.

Fine,[5] when investigating the bonding of pressure welds, developed an etching technique which would detect what were believed to be sub-microscopic voids along the bonding plane of defective welds. The

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etching process was an electrolytic one, using 85% orthophosphoric acid as electrolyte, a stainless-steel plate as cathode, and the specimen to be etched as the anode, with a current density of 1 amp./sq. in. and an applied potential of 4 V. With these conditions, satisfactory results were obtained after etching for 15 min.

Later, Fine[6] reported that the phosphoric acid etch would detect the presence of overheating in steel. However, from his published results, the temperatures he employed for pretreatment of his steels were far in excess of those required to produce overheating, and would most certainly have resulted in burning. Fine[5] stated in his original paper that the phosphoric acid, besides revealing the sub-microscopic voids, also dissolved away non-metallic inclusions. It is not surprising, therefore, that the microstructures developed by this reagent on the steels he examined showed clearly defined pits around the boundaries of the austenite grains. These correspond simply to the non-metallic matter which is invariably present at the boundaries of burnt steel. It has been found by the present authors that the phosphoric acid etch is not suitable for detecting overheated steel.

Hinde[7] reported a comprehensive study of the action of the nitro-sulphuric acid etch, and also developed another etching procedure which could be used to detect overheated steel. This was an electrolytic etch using 2% sulphuric acid as electrolyte and a 10-V., A.C., power supply. This reagent gave a similar type of microstructure to that obtained with the 10% nitro-sulphuric acid etch.

The authors[3] have shown that the fall in mechanical properties produced by high-temperature treatment depends upon the temperature attained in excess of that required to produce initial signs of overheating. The ideal etching reagent for detecting overheating should, therefore, be able to distinguish between incipiently overheated steel, where only a slight decline in the mechanical properties has occurred, and severely overheated steel, where more serious deterioration of these properties is to be expected. Hinde[7] claimed that the indications given by the nitro-sulphuric acid etch could be used to assess the degree of overheating. He based his judgment on the clarity of the etch and the thickness of the boundaries outlined. While the procedure may have some practical application, it is open to criticism. The clarity of the etch is dependent upon the degree of polish attained after etching ; it is also dependent, but to a lesser degree, upon the type of steel used. From a series of photomicrographs, prepared by Frith,[8] of the various degrees of overheating suggested by Hinde, it is very difficult to observe any significant difference between the proposed grades.

In the opinion of the authors, none of the reagents so far suggested for the detection of overheating may be used quantitatively for assessing the degree of overheating.

BURNING

The detection of burning in steel is not a difficult problem. The reasons for this become clear when the nature of the changes which occur during the process are considered.

Treatment of steel at burning temperatures results in incipient melting at the austenite grain boundaries, and the liquid phase formed is enriched particularly with respect to sulphur and phosphorus. Non-metallic inclusions are taken into solution and subsequently reprecipitated around the austenite grains. Voids and intercrystalline separation may occur as a result of flowing of the molten metal or mechanical deformation at the high temperature.

As would be expected from these facts, evidence of burning in steel may be obtained by microscopical examination of a polished but unetched surface. A clearer indication is obtained, however, by examination of the polished specimen after etching with any reagent which is sensitive to the segregation of phosphorus. Many etchants have been suggested which are very satisfactory for detecting burnt steel, and in view of their multiplicity attention will be confined chiefly to the action on burnt steel of those reagents which are also useful in detecting overheated steel.

Many workers have noted that etching steels in nitro-sulphuric acids after treatment at very high temperatures resulted in a reversal of the action of the etchant. Instead of attacking the boundaries in preference to the body of the grain, as in overheated steel, the etchant attacked the grains in preference to the boundaries, and a microstructure showing white boundaries on a black background was obtained (Fig. 3). This reversal of action corresponds to the change from the overheated to the burnt condition.

A similar reversal was noted[3] when etching burnt steel with a saturated solution of ammonium nitrate. The boundaries of the pre-existing austenite grains were blackened, so giving the type of microstructure shown in Fig. 4.

Experiments have shown that ammonium nitrate solution is an ideal reagent for detecting burning, as it is extremely sensitive to this condition. It is superior to the nitro-sulphuric acid etch, for the structure developed by the latter reagent is dependent upon the degree of polish attained after the initial etching treatment.

The use of nital in the examination of burnt steel has been reported by many workers. The usual type of microstructure obtained was a system of light etching boundaries showing regions where partial melting had occurred (Fig. 5).

The reagents developed by Stead[9] and Oberhoffer[10] for showing phosphorus segregation may be used for detecting burnt steel. These reagents do not differentiate, however, between overheated and not overheated steel. The type of structure obtained after etching a burnt steel with Oberhoffer's reagent is shown in Fig. 6.

Fine's reagent, which has been suggested for detecting overheated steel, is really suitable only for detecting burnt steel. The inclusions present at the grain boundaries are dissolved away and the cavities left behind are slightly enlarged. Any sub-microscopic voids at the grain boundaries will be revealed also, giving a type of microstructure shown in Fig. 7.

A summary of the etching characteristics of overheated and burnt steel is given in Table I.

Of all the reagents reviewed, perhaps the most useful, from a practical point of view, as a means of detecting overheating and burning is the ammonium nitrate etchant, using etching conditions developed in a later section of this paper to give the maximum sensitivity. Where it would be impossible to use the electrolytic etching technique, the nitro-sulphuric acid etchant should be used.

These etchants may be used to detect overheating and burning in samples taken from forgings and drop-stampings immediately after fabrication. When the results of the etch tests are doubtful, a fracture test taken after the steel has been fully heat-treated will certainly decide the issue. But where the indications of these etch tests are not definite, the degree of overheating, if this defect is present at all, will be slight, and the fall in mechanical properties only small. Any steels showing evidence of burning should certainly be discarded.

EXPERIMENTAL INVESTIGATION ON THE ACTION OF THE AMMONIUM NITRATE ETCHANT

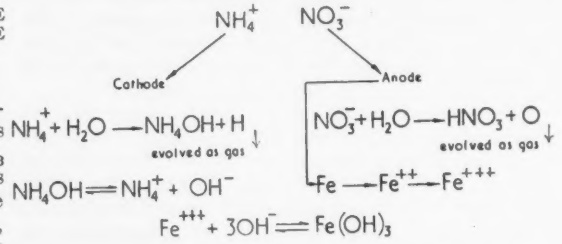
Action on Overheated Steel

Preliminary Investigations—With the etching conditions initially adopted, the NH_4NO_3 etchant was very selective in its action. Using saturated NH_4NO_3 solution as electrolyte, a stainless-steel plate as cathode, and making the specimen to be etched the anode, with a current density of 0·2 amp./sq. cm., white boundaries could not be produced in steel after air-cooling from above the overheating temperature. Neither were they obtained very clearly after quenching from above the Ar_3 point. Toughening the steel by tempering at 600° C. produced a condition most sensitive to the action of the reagent. A further difficulty was that, even in the toughened condition, different types of alloy steels differed in their response to the action of the etchant.

In spite of these difficulties the indications given by the NH_4NO_3 were much clearer than those given by the nitro-sulphuric acids, when used on suitable steels in the fully heat-treated condition. In view of these facts this investigation has been carried out with the object of making the NH_4NO_3 etchant more universally applicable and gaining further information upon its action in differentiating between overheated and not overheated steel.

It was noted[3] that during etching gases were evolved at the anode and cathode ; these were identified as oxygen and hydrogen respectively. As etching progressed the solution turned brown and eventually ferric hydroxide was precipitated. The *p*H of the whole electrolyte was measured while these changes were taking place, and it was found that the *p*H increased rapidly from an initial value of 4·3 to 7·0, after which it remained constant at this value over a prolonged period.

These facts suggest that the reactions occurring are :



Potential Drop at the Anode/Electrolyte Interface—In order to obtain further information on the etching process, an examination was made of the potential drop at the anode/electrolyte interface. The apparatus used for this purpose is shown in Fig. 9. It consisted of a saturated calomel electrode, the tip of which was placed close to the anode surface. The total current flowing was measured by the milliammeter *A*, and the total potential drop by the voltmeter

Table I

ETCHING CHARACTERISTICS OF OVERHEATED AND BURNT STEEL

Reagent	Method	Action on Overheated Steel	Action on Burnt Steel
2½% nitric acid in ethyl alcohol	Swab surface for 30 sec.	May produce grain contrast but not indicative of overheating	White boundaries outlining pre-existing austenite grains
Saturated aqueous solution of ammonium nitrate	Electrolytic, specimen anode. Current density 1·0 amp./sq. cm.	White boundaries outlining pre-existing grains	Black boundaries outlining pre-existing austenite grains
10% nitric acid 10% sulphuric acid in water	Etch for 30 sec., swab surface. Repeat three times and then repolish lightly	Black boundaries outlining pre-existing austenite grains	White boundaries outlining pre-existing austenite grains
85% orthophosphoric acid	Electrolytic. Specimen anode. Current density 0·15 amp./sq. cm. Etching time 15 min.	Does not differentiate between overheated and not overheated steel	Attacks inclusions at the grain boundaries
Oberhoffer's reagent	Swab surface for 30 sec.	Does not differentiate between overheated and not overheated steel	Shows phosphorus segregation at grain boundaries

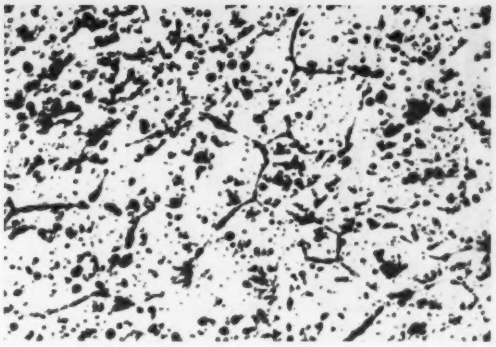


Fig. 1—Overheated steel etched in 10% nitro-sulphuric acid, showing black boundaries × 50

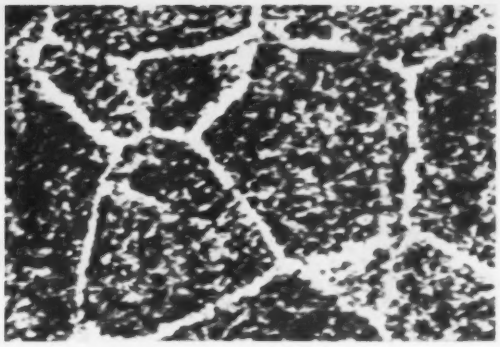


Fig. 2—Overheated steel etched electrolytically in saturated ammonium nitrate solution, showing white boundaries × 50

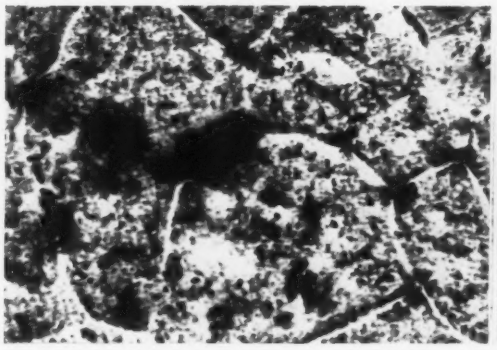


Fig. 3—Burnt steel etched in 10% nitro-sulphuric acid, showing white boundaries × 50

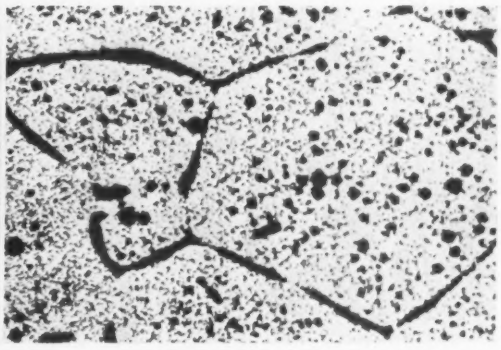


Fig. 4—Burnt steel etched electrolytically in saturated ammonium nitrate solution, showing black boundaries × 50

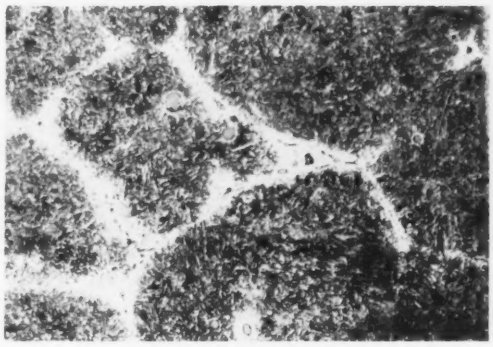


Fig. 5—Burnt steel etched in 2½% nital, showing white boundaries and non-metallic inclusions at the boundaries × 50

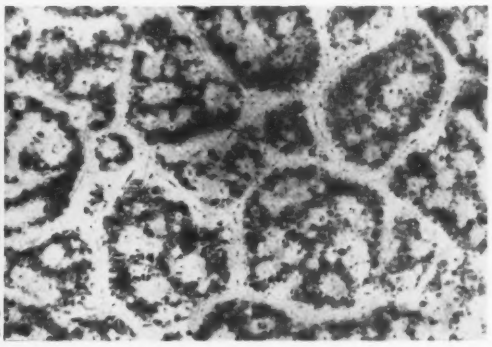


Fig. 6—Burnt steel etched in Oberhoffer's reagent, showing phosphorus segregation × 50

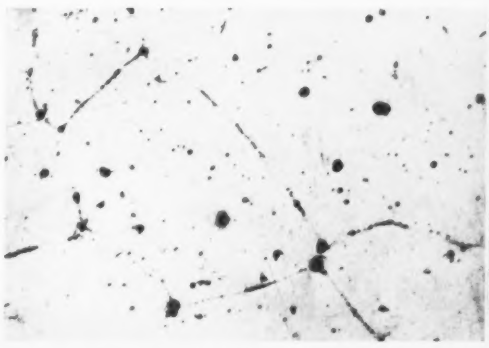


Fig 7—Burnt steel etched in Fine's reagent (85% orthophosphoric acid), showing enlarged voids × 25

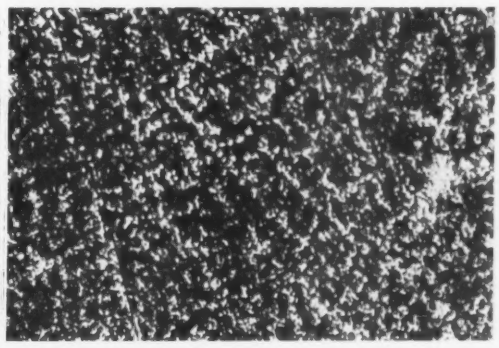


Fig. 8—Nickel steel in the wrought condition, etched electrolytically in saturated ammonium nitrate solution, showing dendritic structure

V. The potential drop across the interface was measured potentiometrically. The surface to be etched was prepared by mechanically polishing in the normal manner, and then the specimen was suspended in the electrolyte on platinum wire. Reaction was confined to the polished surface by covering the remaining material with paraffin wax.

Effect of Varying the Applied E.M.F.—The effect of the applied e.m.f. upon the current flowing in the circuit and the effect of the anode current density upon the potential drop across the anode/electrolyte interface were investigated, and the results are shown in Fig. 10. In carrying out the experiments, the electrolyte was first ' conditioned ' by etching a control specimen until ferric hydroxide was precipitated. The test specimen was then placed in position and measurements were taken at increasing current densities. With this technique reproducible results were obtained. No marked variations of the anode/electrolyte interfacial potential with time at constant current density were observed over a period of 30 min.

Effect of Varying the Current Density—Etching experiments were carried out at various current densities. In this series the specimen was repolished after each etching test. The steel used was one which previous experiments had shown to be responsive, when overheated, to the action of the NH_4NO_3 after etching at a current density of 0·2 amp./sq. cm. It was found that before white boundaries were produced a certain current density had to be exceeded ; increasing the current density beyond this critical value gave better differentiation between the boundary and the grain, until at the maximum current density available with the apparatus very clear boundaries were produced after etching for 30 sec.

Effect of Varying the Concentration of the Electrolyte—The effect of the concentration of the electrolyte was next investigated. It was found that the greater the concentration the higher was the anode/electrolyte potential drop for any given current density. The results of these experiments are shown in Fig. 11. Etching experiments indicated the minimum current

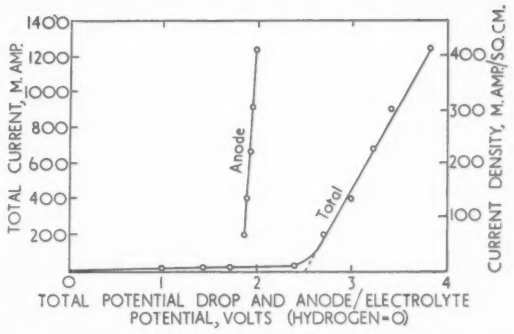


Fig. 10—Effect of applied e.m.f. upon the current and anode potential. Decomposition potential 2·45 V.

density required to produce boundaries. These showed that at the highest current density available with the apparatus it was impossible to produce white boundaries with a 6·25% saturated solution. With a 25% saturated solution boundaries could be produced, but a higher current density was required than with a saturated solution.

Effect of Different Electrolytes—Further experiments were undertaken using different electrolytes. In the first series the acid radical was changed, the ammonium radical being retained. In each case a saturated solution of the salt was used. It was found that chloride, sulphate, and persulphate were ineffective in producing white boundaries. In similar experiments, in which the nitrate radical was retained, fairly clear white boundaries were obtained with the potassium and sodium salts after etching at the highest attainable current density. These experiments suggest that the concentration of NO_3^- is the important factor and, as the ammonium salt is more soluble than the potassium or sodium salt, it is a more effective etchant.

In view of the fact, already noted, that the higher the current density the better the differentiation between the boundary and the grain, etching tests

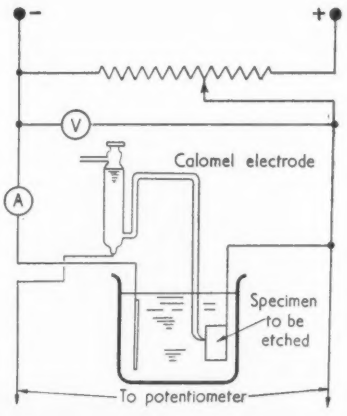


Fig. 9—Apparatus for measuring electrode potential

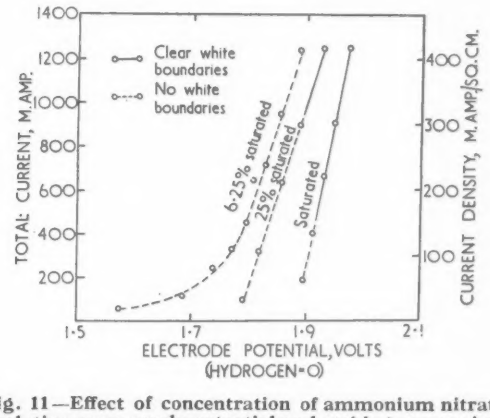


Fig. 11—Effect of concentration of ammonium nitrate solution upon anode potential and etching properties

were carried out on steels in differing conditions of heat-treatment. Using a current density of 0·2 amp./sq. cm. and saturated NH_4NO_3 solution as electrolyte, no white boundaries were found in severely overheated steel in the as-cooled condition. By increasing the current density to 1·0 amp./sq. cm. clear white boundaries were produced even in steels only incipiently overheated; but it was impossible to produce this effect in steels which did not show evidence of overheating in the fracture test. Similar results were obtained when etching overheated steels after oil-quenching from above the Ar_3 point.

Tests were then carried out on steels in the fully heat-treated condition which, after etching with a current density of 0·2 amp./sq. cm., were not responsive to the action of the NH_4NO_3. It was found that these steels showed clear white boundaries after overheating when etched electrolytically in saturated NH_4NO_3 solution at a current density of 1·0 amp./sq. cm. This applied in all conditions of heat-treatment.

These experiments have removed, therefore, many of the difficulties encountered initially when using this reagent as a means of detecting overheating in steel.

CONSIDERATION OF RESULTS

The action of the etchant in differentiating between overheated and not overheated steel is difficult to interpret. In the overheated condition the boundary system obtained after treatment at high temperatures retains a difference in composition. The behaviour of the nitro-sulphuric acid in attacking preferentially the boundaries of the pre-existing austenite grains indicates that the change in composition results in the boundaries behaving as though they were more electronegative than the remainder of the grain. This view is difficult to reconcile with the behaviour of the NH_4NO_3 etchant in leaving the boundaries unattacked. However, in view of the high concentration of NO_3^- it is probable that passivity of the anode must ultimately occur. Preferential passivity of the boundary material produced by overheating would result in a system of lightly attacked boundaries and more deeply attacked grains. The view that passivity has an important influence is supported by the fact that the higher the NO_3^- concentration the lower the current density required to produce white boundaries, and the fact that Cl^-, SO_4^{--}, and $S_2O_8^{--}$ are not effective in producing white boundaries.

The characteristics of the NH_4NO_3 etchant when applied to burnt steels were investigated in a manner similar to that adopted when studying overheated steels. Black boundaries were produced in burnt steel over a wide range of current densities, solution strengths, and etching times. The effect was also independent of the condition of heat-treatment of the steel.

With burnt steels the NH_4NO_3 preferentially attacked the grain boundaries, which suggests that the change in composition produced by burning is entirely different from that obtained by overheating. The white boundaries obtained with nitro-sulphuric acid etch on steels in the burnt condition suggest that the change in composition has resulted in the boundary material becoming more electropositive than the remainder of the grain. At the same time the boundary material may be made passive less readily than the grain when attacked by NO_3^-.

The preferential attack at the austenite grain boundaries of burnt steel by the NH_4NO_3 etchant, when it was known that phosphorus segregation at the boundaries must have occurred, suggested that this etchant may be useful in detecting segregation in cast and wrought steels. This was investigated by Lucas,[11] and Fig. 8 is a photomicrograph of a 4½% nickel steel in the wrought condition after electrolytically etching in NH_4NO_3 solution at a current density of 1·0 amp./sq. cm. The phosphorus segregation was clearly indicated and, as would be expected, the phosphorus-rich interdendritic material etched black.

These experiments indicate that electrolytic etching processes deserve more attention than they have yet received. It should be possible to develop electrolytic etching methods which will indicate the segregation known to exist in aluminium–copper alloys from the evidence of microradiographs, and in many other alloys where the normal etching reagents indicate almost complete homogeneous solid solutions.

Acknowledgments

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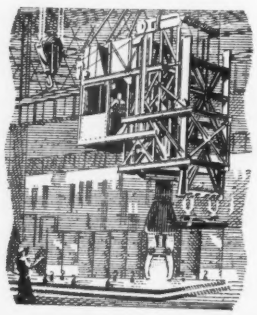

Grain-Boundary Phenomena in Severely Heated Steel

By T. Ko, B.Sc., Ph.D., and D. Hanson, D.Sc.

SYNOPSIS

A study has been made of the structure and tensile properties of carbon steels which have been quench-hardened after being heated to very high temperatures in the austenitic range. A new metallographic method of studying overheating has been used which shows that, on overheating, sulphide inclusions are precipitated at the austenite grain boundaries ; these inclusions cannot be seen in an ordinary microsection. A subsidiary investigation of some specially prepared iron alloys has shown that the presence of manganese sulphide is necessary for the precipitate to occur. Two effects are found when the steel has been heated to temperatures very near the solidus : the sulphides are precipitated at the grain boundaries in the form of a eutectic, and there is a segregation of phosphorus to the precipitated regions.

The effect of high austenitizing temperatures on the tensile properties of martensite has been studied on low-carbon steels and on an Fe–C–Mn alloy. Martensitic plain carbon steels containing 0·08, 0·17, and 0·23% of carbon have tensile strengths of 73, 95, and 110 tons/sq. in. respectively, and have reductions in area of 40–50% when fine-grained. The ductility decreases with increasing austenitizing temperature ; in steels heated below the overheating temperature this loss of ductility can be recovered by a single normalizing treatment, but in overheated steels little recovery can be achieved except by subjecting the material to rather elaborate heat-treatment.

The theory is put forward that overheating is due to the partial precipitation on the austenite grain boundaries, at high temperatures, of sulphide brought into solution during heating ; the precipitation takes place when cooling, at a suitable rate, to the overheating temperature. Segregation of phosphorus occurs when the solidus/liquidus range of the steel (the solidus being lowered by the presence of sulphur in solution) is entered at very high temperatures.

Introduction

THE effect of high-temperature treatment, between the upper critical point and the solidus, on the structure and the mechanical properties of steel has received considerable attention since the pioneer work of Stansfield[1] and Stead,[2] particularly because of its importance in welding and forging.

It is now generally recognized that lack of ductility of the hardened zone next to a weld may result in the occurrence of underbead cracking in the welding of certain steels.[3, 4] One of the factors which can affect the ductility is the thermal history of the steel : the ductility of hardened steel generally decreases with increasing austenitizing temperature, and becomes very low when the steel has been heated to temperatures near the solidus.[5] Part of this loss of ductility is persistent, *i.e.*, it cannot be recovered by a *single* normalizing treatment, and is accompanied by the appearance of facets in the fracture when the steel is suitably heat-treated, and of a cellular structure when it is properly etched with specific chemical reagents. Steel in such a condition is said to be overheated, and the temperature above which the phenomenon occurs is the overheating temperature. On the other hand, a steel is said to be burnt when incipient fusion of the metal occurs at very high temperatures.

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Overheating and burning have received close attention from many investigators,[6, 7, 8] and recent work by Preece *et al.*[9] and others[10, 11] has contributed much towards the understanding of the phenomena.

It is now generally accepted that :

(i) Overheating of a steel occurs when it is heated above a certain temperature, which varies from steel to steel[7, 9, 10]

(ii) The overheating temperature is an inherent property of the steel, and is not affected by the furnace atmosphere[9, 10]

(iii) Overheating is revealed by the presence of facets in the fracture of a suitably heat-treated specimen and of a boundary network after suitable etching[7, 9, 10]

(iv) Overheating results from changes at grain boundaries, which are retained to a certain extent after heat-treatment ; these lead to preferential attack by chemical reagents, and introduce a weakness in the structure[6, 7, 9–12]

(v) Overheating may be suppressed by very slowly cooling the steel through the overheating range[9, 10]

(vi) Working the overheated steel reduces the effect of overheating[9, 10]

(vii) Overheated steels can be restored by one of the following treatments : (*a*) reheating to a high temperature and cooling very slowly through the overheating range[9] ; (*b*) repeated normalizing and long annealing ; (*c*) repeated normalizing at successively lower temperatures.[12]

EXPERIMENTAL

Study of previous work suggested that the best approach to the problem would be to examine the nature of the austenite grain boundaries in overheated specimens. However, examination of the boundaries in an ordinary microsection is unsatisfactory because the grain-boundary change produced by overheating is either too fine to be seen or is obscured by the distortion of the surface during polishing ; even electrolytic polishing does not avoid this

difficulty because it causes localized attack on the inclusions.

A metallographic study of the grain boundaries was made possible, however, by the observation that, when an overheated and hardened steel was fractured, many intercrystalline facets were produced on the fracture, and these facets were sufficiently plane to be examined directly under the microscope, thus making possible the direct observation of the austenite grain surface.

Examination of these facets revealed that numerous inclusions had deposited on their surfaces (Fig. 3). These observations showed that experimental evidence of grain-boundary precipitation during overheating could be obtained, and the effect of temperature and others factors on the phenomenon could be studied if the intercrystalline surface of austenitic grains of an overheated steel could be exposed.

It is well known that when high-carbon steel is severely quenched intercrystalline cracking occurs, owing to stresses set up by the martensite transformation. In the present work, therefore, steels of high hardenability were heated to high temperatures, cooled at a suitable rate to a temperature below the overheating temperature, quenched to the martensitic state and fractured to expose the original austenite grain surfaces, which were then studied microscopically.

In conjunction with this study of grain interface precipitation of inclusions, the effects of high-temperature heating on the mechanical properties of martensitic low-carbon steel and on an Fe–C–Mn alloy have also been examined.

MATERIALS USED

Five plain carbon steels, one Ni–Cr–Mo steel, one free-cutting steel, and five alloys prepared from spectrographically pure materials were used; the analyses are given in Table I.

The alloys, *F*1 to *F*5, were made from Johnson Matthey H.S. Iron 846, H.S. graphite J.M.3, electrolytic manganese (99% of Mn), chemically pure sulphur, and phosphorus pentoxide. The iron used, according to the supplier, contained up to 0·0003% of sulphur and spectrographic traces of copper, nickel, cobalt, and silicon. The iron used in making *F*5 contained also traces of manganese.

Individual charges of about 50–100 g. each were melted by H.F. induction heating in a pure alumina crucible. The charge was melted under dried argon and cast in the crucible. The ingot, unless otherwise stated, was homogenized at 950° C., normalized, heated by induction to about 1000° C. and forged to $\frac{1}{2}$ in. × $\frac{1}{4}$ in. flats, homogenized again at 950° C. in argon for 6 hr., and finally forged to bars of $\frac{9}{32}$ in. dia.

FURNACE EQUIPMENT

The furnace used for heating the specimens was modified from the equipment used by Wheeler and Kondic,[5] to allow the rapid quenching of a specimen from high temperature without exposing it to air.

The furnace tube *T*, Fig. 1, is of mullite, 36 in. long and $\frac{7}{8}$ in. bore. The Kanthal-wound furnace *K* is used for preheating the specimen and for cooling it to the quenching temperature; temperature regulation is by means of barretter control lamps. *P* is a high-temperature furnace consisting of a platinum winding *R* on the furnace tube *T*, supplied from a continuously variable tapping transformer, and an outer auxiliary Kanthal winding on a concentric alumina tube *Q*.

Connected to the top of furnace tube *T* is a water-cooled chamber from which the furnace tube can be evacuated and refilled with argon through different valves. The specimen is suspended on a 0·1-mm. tungsten wire from the pulley *D*; this wire passes through a small hole in the cap *A*, sealed by a rubber gland *B*, which can be tightened or relaxed by the screwed cap *A*.

The lower end of the tube *T* is connected to a flanged steel tube *V* by a rubber band *U*. A mullite thermocouple sheath is sealed in through an opening on one side of the tube *V*. The flanged bottom is covered with a ground-steel plate *W* which can be forced open with

Table I

ANALYSES OF MATERIALS USED

Steel No.	C	Si	Mn	P	S	Ni	Cr	Mo	Cu	Al	N	O	Method of Manufacture
*M**	0·08	0·21	0·53	0·034	0·031	0·08	0·02	0·01	0·05	0·0031	0·010	0·017	H.F. induction
N	0·17	0·20	0·59	0·036	0·029	0·08	0·02	0·02	0·045	0·0049	0·011	0·006	H.F. induction
P	0·23	0·21	0·56	0·038	0·029	0·08	0·02	0·01	0·045	0·0049	0·010	0·011	H.F. induction
S	0·47	0·21	0·62	0·034	0·028	0·08	0·02	0·01	0·045	0·0036	0·009	0·012	H.F. induction
T	0·85	0·20	0·64	0·034	0·027	0·09	0·02	0·01	0·05	0·0045	0·014	0·003	H.F. induction
VB	0·32	0·22	0·63	0·023	0·028	0·60	0·51	0·55		0·0045	0·009		Basic electric
HS	0·14	trace	1·01	0·13	0·24								Unknown
*F*1	0·48												H.F. induction
*F*2	0·48		0·55		0·049					*			H.F. induction
*F*3	0·56			0·089									H.F. induction
*F*4	0·24												H.F. induction
*F*5	0·19		0·52										H.F. induction

* Determination of aluminium made to detect possible contamination of the alloy by the refractories was found to be 0·006%

a screw. To quench the specimen, the tungsten wire is cut above the cap *A*, the quenching bath *Y* is lifted until the flange is submerged below the surface of the quenching medium, the cover plate *W* is forced open, and the specimen is immediately dropped into the quenching tank by relaxing the rubber gland *B*.

Before heat-treatment the furnace tube is usually evacuated to 0·1 mm. of mercury and refilled with argon dried with phosphorus pentoxide. These precautions in atmosphere control made it possible (*a*) to obtain specimens with perfectly bright surfaces, without appreciable oxidation or decarburization after heating and quenching, and (*b*) to fully harden 0·08% plain carbon steel (in the form of tensile specimens of $\frac{1}{8}$ in. dia.) in agitated 20% brine, to a maximum tensile strength of 73 tons/sq. in. and a V.D.P. hardness of 360. Tarnishing of the specimen due to an imperfect vacuum, or to a delay between opening of the bottom plate and quenching, or to replacing the brine with tap water, inevitably

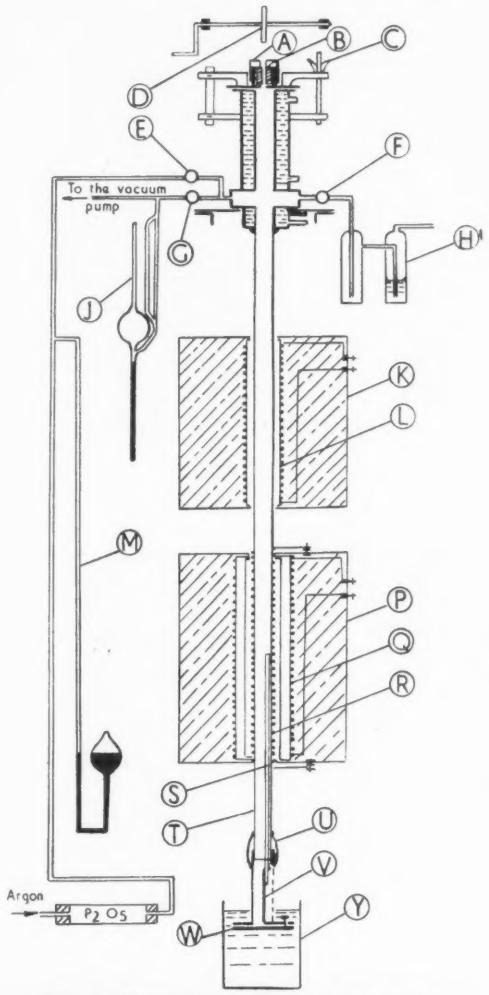


Fig. 1—Furnace used for heat-treatment

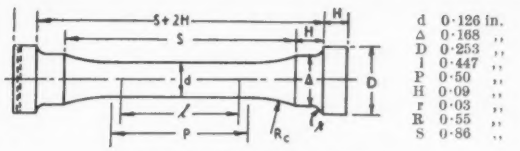


Fig. 2—Tensile specimen

gives specimens of much lower strength, because of precipitation of ferrite at grain boundaries, even in the case of steel *P* of 0·23% of carbon. This is shown in Fig. 4, taken from a specimen of steel *P* heated to 1440° C. and quenched in tap water at 21° C.

EXPERIMENTAL PROCEDURE

Study of Grain-Boundary Precipitation

Specimens for investigation of grain-boundary precipitation, $\frac{1}{4}$ in. dia. and 1 in. long, were cut longitudinally from rolled bar, and were notched at $\frac{1}{4}$ in. from one end. Each specimen, suspended on tungsten wire, was admitted into the water-cooled chamber. After introducing argon, the specimen was lowered into the low-temperature furnace *K*, at 950° C., where it remained for 10 min., and was then transferred to the high-temperature furnace *P* for 10 min. After this, the specimen, unless otherwise stated, was returned to the low-temperature furnace for 5 min., and quenched in brine or still tap water. The specimen was then fractured at the notch, and the intercrystalline facets were immediately examined. That the fracture facets so obtained correspond to the original austenite grain boundaries is evident from Fig. 5, taken from a specimen of steel *S* heated to 1370° C.; both the fracture and the cracks are intercrystalline.

The specimen was then prepared for metallographic study in the conventional manner.

Study of Mechanical Properties

For the study of the effect of the initial heating temperature on the mechanical properties of fully hardened steel, tensile specimens, shown in Fig. 2, machined longitudinally from rolled bars and degreased, were heat-treated in the equipment described on p. 52. Each specimen was first treated in the furnace *K* at 950° C. for 10 min. This allowed all specimens to transform to austenite under identical conditions, and removed the uncertain effect of the initial grain size (which increases with decreasing heating rate through the critical range) that would result if specimens were directly heated in a high-temperature furnace maintained at different temperatures for different specimens. After the preliminary heating, each specimen was lowered into the high-temperature furnace for 3 min., thus allowing the attainment of the desired high temperature without the occurrence of excessive grain growth. The specimen was then quickly returned to the low-temperature furnace for 3 min., and finally quenched in 20% brine (series *A*).

In a second series of investigations (series *B*), in which the effect of overheating on tensile properties was studied, each specimen, after being heated to high temperature and cooled to 950° C., was normalized by returning it to the water-cooled chamber for

10 min. It was then reheated at 950° C. for 10 min., and quenched in 20% brine.

All specimens were tested, within 24 hr. of quenching, on a Hounsfield tensometer with an idling crosshead speed of 0·1 in./min.

EXPERIMENTAL RESULTS

GRAIN-BOUNDARY PRECIPITATION

Grain-boundary precipitation in steels *T*, *S*, *VB*, and alloys *F*1 to *F*3 was examined for the temperature range between 950° C. and the solidus. In steels *P*, *N*, and *M*, it was possible to obtain intercrystalline fractures only when the steels were severely overheated.

Steels *T*, *S*, *VB*, *P*, and *N*

When the steels were quenched from 950° C. without being previously heated to very high temperatures, the surfaces of the austenite grains were very clean (Fig. 6); a few inclusions could be seen, which seemed to extend above and below the surface.

On the other hand, heating to high temperatures caused inclusions to appear on the intercrystalline surfaces. Three stages were observed:

(1) The appearance of very fine inclusions without any change in microstructure

(2) The precipitation of fine inclusions of Mn-rich sulphide in sufficient amount to cause the appearance of the overheating structure when the specimen was etched with nitro-sulphuric acid

(3) The formation of Fe-rich sulphide, which appeared as one component of a eutectic and could be seen on a polished section. This change was accompanied by the metallographically visible segregation of impurities at the grain-boundary regions, which increased the resistance of the steel to tempering, and gave a white-etching network when the specimen was etched with nital.

The approximate temperature ranges in which the different stages of precipitation were observed are summarized below. The transition from stage 1 to stage 2 is a gradual one; the lower temperature limits of stage 2 are those at which the specimens were revealed as overheated when etched with nitro-sulphuric acid.

Steel	Stage 1, °C.	Stage 2, °C.	Stage 3, °C.
T	1150–1300	1300–1350	Above 1350
S	1250–1350	1350–1420	Above 1420
VB	1300–1400	1400–1430	Above 1430

The fine inclusions present on the intercrystalline surface in stage 1 are illustrated in Fig. 7. The amount of the fine inclusions could be reduced by quenching the specimen directly from the high temperature; but both the number and the size of the inclusions increased with increasing temperature. That they decreased in amount with increasing cooling rate indicates that these inclusions were precipitated during cooling. On the other hand, had they been formed during heating, rapid quenching from the high temperature would have prevented their solution during cooling and increased their amount on the grain interface. The microstructure at this stage showed an ordinary martensitic structure when etched with nital, and gave no overheating structure when etched with nitro-sulphuric acid.

The inclusions on the intercrystalline surface in stage 2, shown in Fig. 8, were similar in shape to those of stage 1, and were about 0·0005 mm. across at the initial stages. The size, however, increased with increasing temperature. They were very thin in the direction normal to the fracture, for examination of the grain boundaries in a polished and lightly etched section revealed no trace of these inclusions. Orientated inclusions were sometimes observed, *e.g.*, those on the left-hand side of the facet in Fig. 9. The presence of these inclusions in the crystalline form supports the view that they were precipitated as a solid. The amount and size of the inclusions were not affected by prolonged heating at constant temperatures, but the amount could be reduced by quenching the specimen from the high temperature.

The inclusions, grey in colour, soluble in 2% sulphuric acid, attacked by chromic acid, and removed completely by alkaline sodium picrate, are believed to be Mn-rich sulphide. The nitro-sulphuric acid etch showed that the specimens had been overheated (Fig. 10); the intensity of overheating increased with increasing temperature. Preece and his co-workers[9] have found a saturated aqueous solution of ammonium nitrate to be a satisfactory etchant for the detection of overheating. Accordingly, the specimens of steels *T* and *VB* were reheated after quenching to 950° C. for 30 min., oil quenched, and tempered for 1 hr. at 200° C. (steel *T*) or 600° C. (steel *VB*), and then electrolytically etched with this reagent. As shown by Fig. 11, specimens of steels *T* and *VB* initially heated above 1325° and 1400° C. respectively, were identified by this etch as overheated.

When the steel was heated to higher temperatures, (*e.g.*, above 1350° C. in the case of steel *T*) the number of facets containing the usual grey inclusions of irregular shapes decreased in favour of a second type of facet surface. These facets contained two kinds of inclusions: (1) large, round inclusions which showed a circle of influence within which no other inclusions were present, and (2) either small, round inclusions or inclusions deposited in long fascias closely resembling ferns, or both. These are illustrated in Figs. 3 and 12*a*. The amount of fern inclusions increased with (*a*) increasing carbon content, (*b*) increasing temperature, and (*c*) decreasing cooling rate.

The shape of the fern inclusions varied slightly with the carbon content of the steel, the fascias being radially arranged in the low-carbon steels. Figure 12*b* shows the inclusions in high-carbon steel and Fig. 13*a* shows them in low-carbon steel. Parts of the ferns were sometimes replaced by clusters of fine inclusions arranged in rows, as shown in Figs. 13*b* and 13*c*.

These photomicrographs show that the fascias of the fern inclusions are not usually connected to the ribs and therefore the inclusions could not be formed by dendritic growth. The only possible explanation for the formation of both the ferns and the fine, round inclusions is that they were precipitated as one component of a eutectic, the other being probably iron. The round inclusions which show circles of influence must have separated before the others.

The proportion of large, round inclusions to ferns depends on the rate of cooling from the high

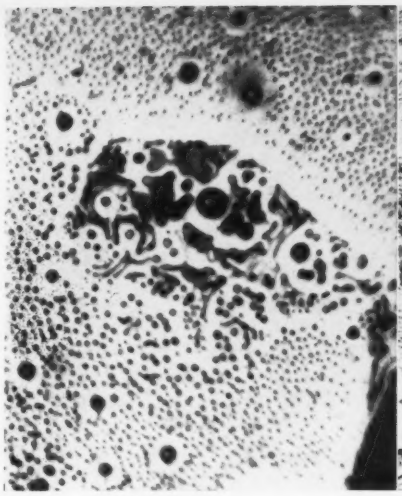


Fig. 3—Inclusions on grain interface in steel *P* heated to 1430° C. × 800

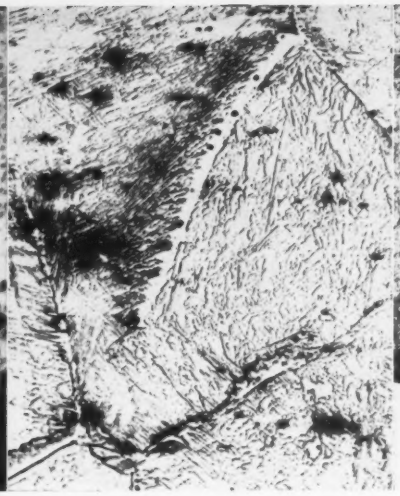


Fig. 4—Ferrite in quenched steel *P* × 400

Fig. 5—Intercrystalline fracture and cracks in steel *S* heated to 1370° C. × 100

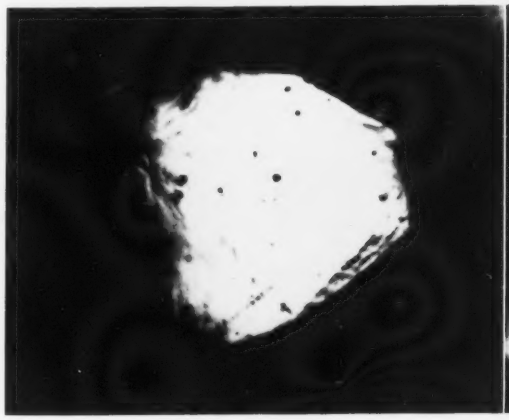
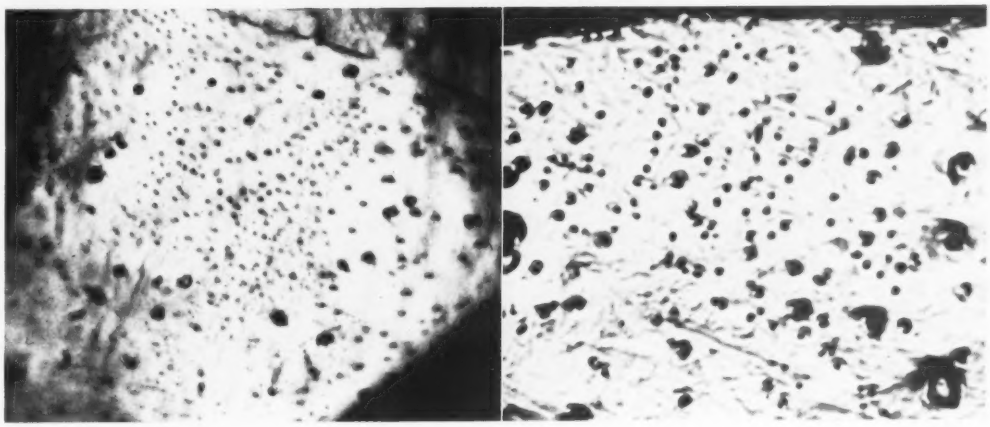


Fig. 6—Grain interface of steel *T* heated to 1050° C. × 800

Fig. 7—Grain interface inclusions in steel *T* heated to 1250° C. × 500



(*a*) × 800 (*b*) × 1200

Fig. 8—Grain interface inclusions in steel *T* heated to (*a*) 1300° C., (*b*) 1325° C.

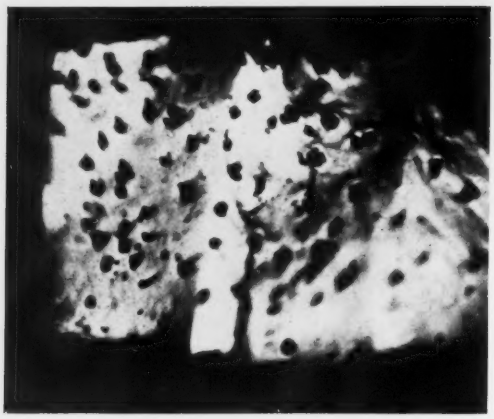


Fig. 9—Grain interface inclusions in steel ***S*** heated to 1440° C. × 800

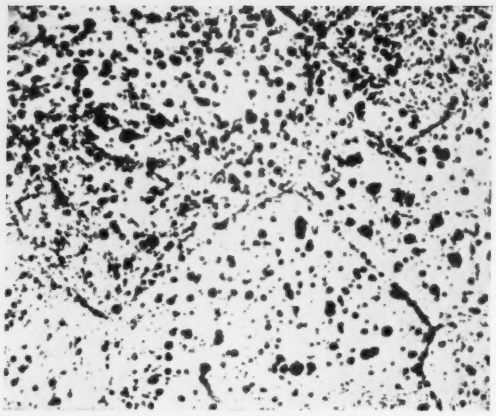
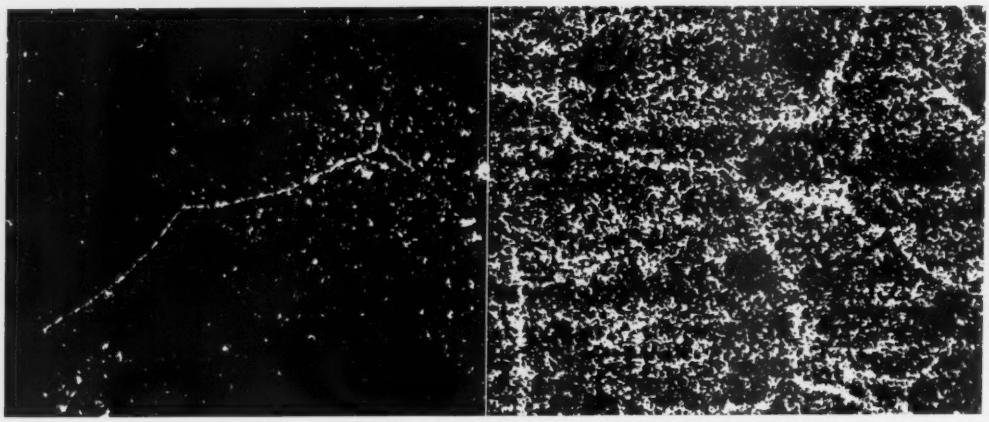
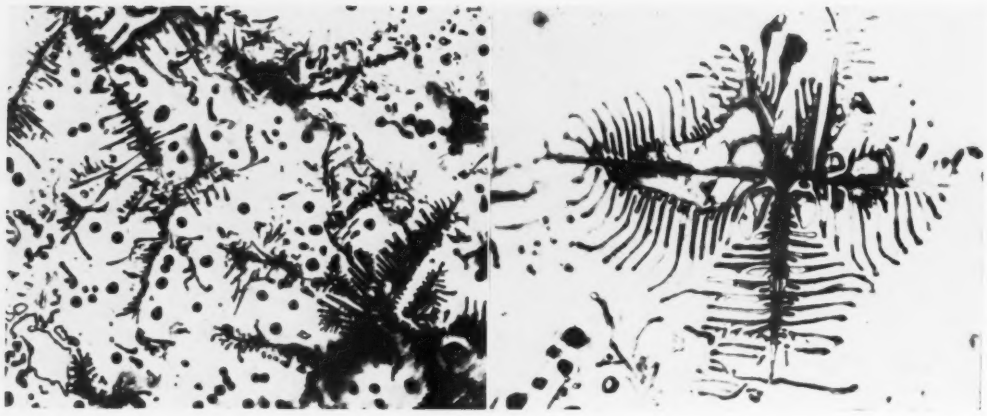


Fig. 10—Steel ***T*** heated to 1300° C. Etched with nitro-sulphuric acid × 150



(*a*) × 100 (*b*) × 50

Fig. 11—Overheated structures (*a*) steel ***T*** heated above 1325° C. tempered at 200° C. for 1 hr., (*b*) steel ***VB*** heated above 1400° C. tempered at 600° C. for 1 hr. Etched with ammonium nitrate

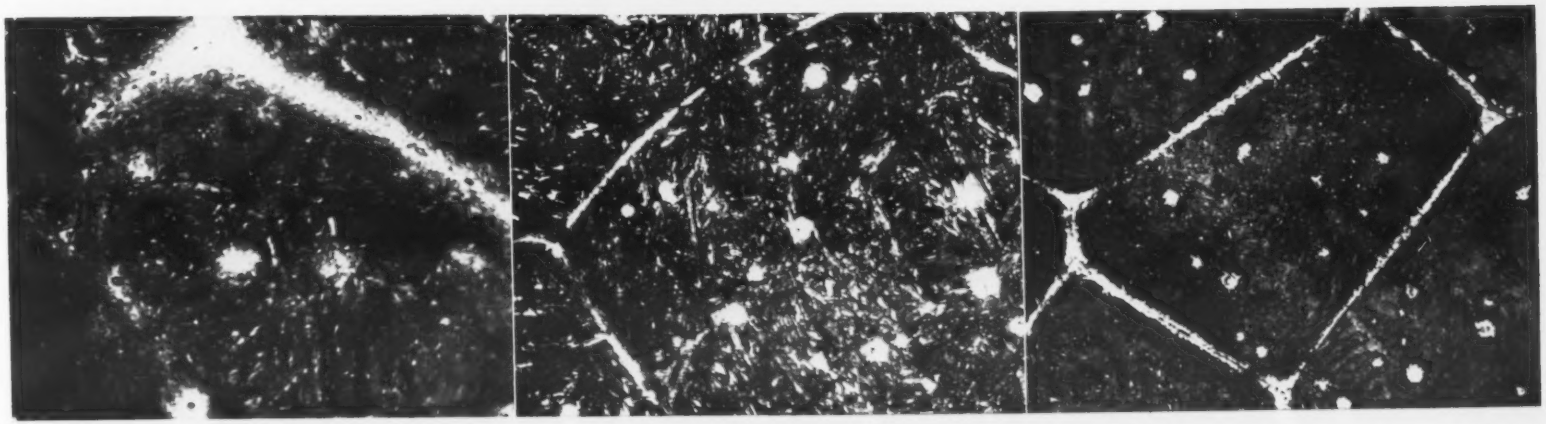


(*a*) × 425 (*b*) × 500

Fig. 12—Grain interface inclusions in steel ***T*** (*a*) quenched from 1360° C., (*b*) heated to 1360° C.

(*a*) × 800 (*b*) × 1200 (*c*) × 1200

Fig. 13—Grain interface inclusions in (*a*) steel *P* heated to 1430° C., (*b*) steel *T* heated to 1360° C., (*c*) steel *VB* heated to 1450° C.



(*a*) × 500 (*b*) × 500 (*c*) × 400

Fig. 14—White-etching areas in martensitic structure in steel *T*, (*a*) heated to 1360° C., quenched from 950° C., tempered. Etched with 2% nital, (*b*) quenched from 1360° C., tempered. Etched with 2% nital, (*c*) quenched from 1360° C. Treated with Stead's reagent

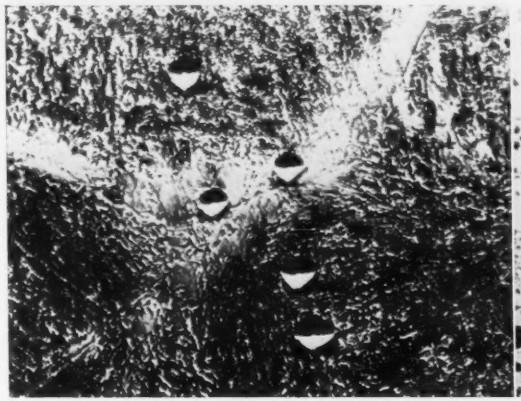
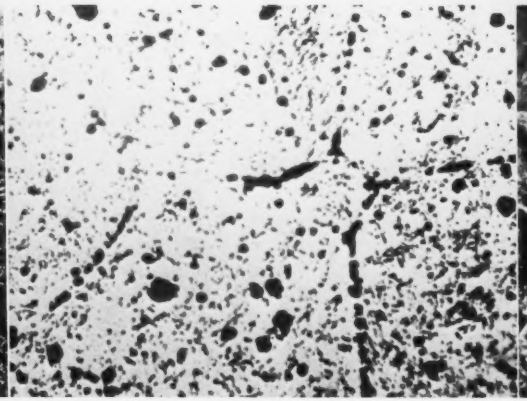
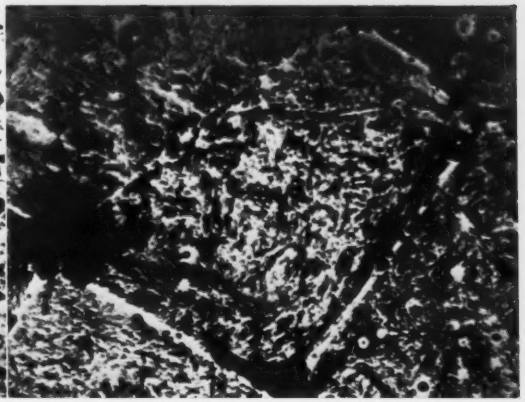


Fig. 15—Microhardness indentations in steel *VB*, air-cooled from 1490° C. × 550



(*a*) × 150



(*b*) × 400

Fig. 16—Structure in steel *T* heated to 1360° C. Etched with nitro-sulphuric acid (*a*) as-quenched, (*b*) after tempering

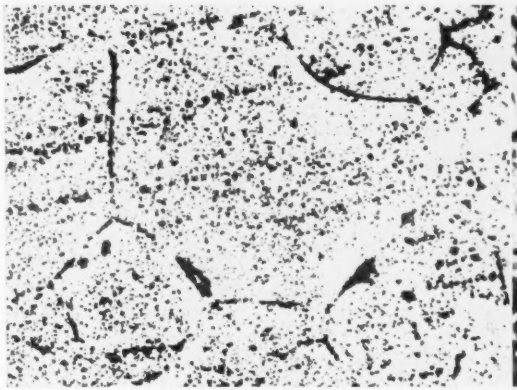


Fig. 17—Structure in steel *VB*, heated to 1440° C., air-cooled, quenched from 950° C., tempered at 600° C. Etched with ammonium nitrate × 50

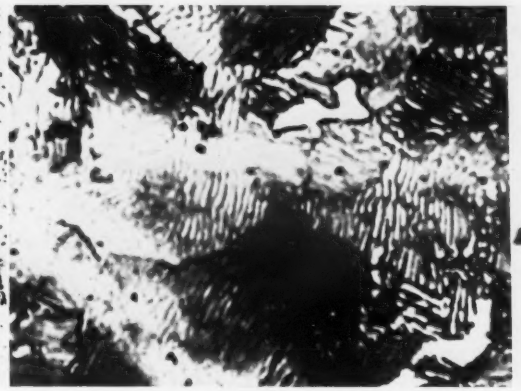


Fig. 18—Steel *T* heated to 1360° C., quenched, reheated to 900° C., air-cooled × 3000

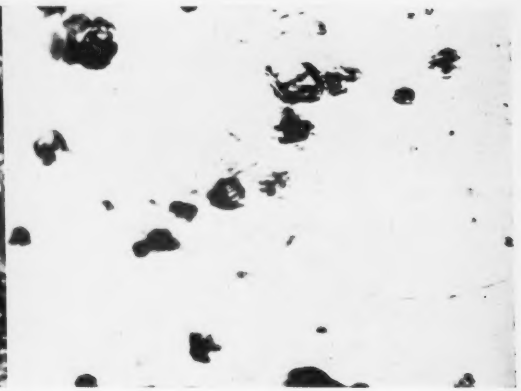
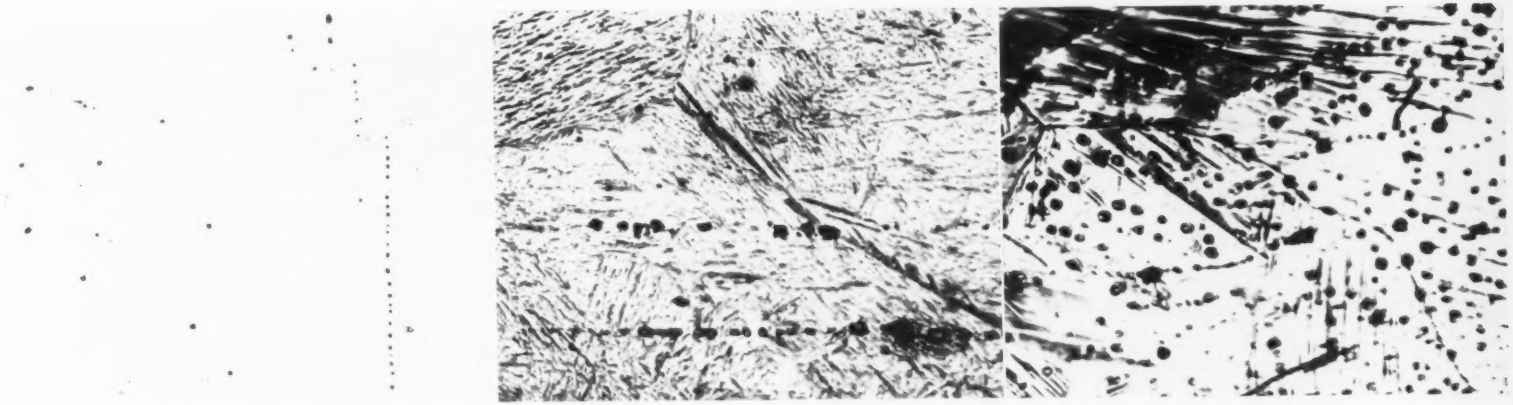


Fig. 19—Grain interface inclusions in steel *T* very slowly cooled from 1360° C. × 500



(*a*) × 500 (*b*) × 500 (*c*) × 500

Fig. 20—Inclusions in steel *T* heated to 1360° C. (*a*) quenched from 1360° C., (*b*) very slowly cooled to 1300° C., quenched from 950° C., (*c*) slowly cooled to 950° C., quenched

Fig. 21—Grain interface inclusions in steel *M* heated to 1490° C., quenched from 950° C. × 500

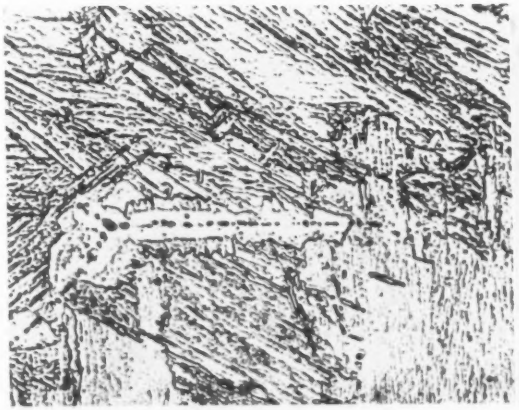


Fig. 22—Transcrystalline inclusion network in steel *M*, quenched from 1490° C. × 500

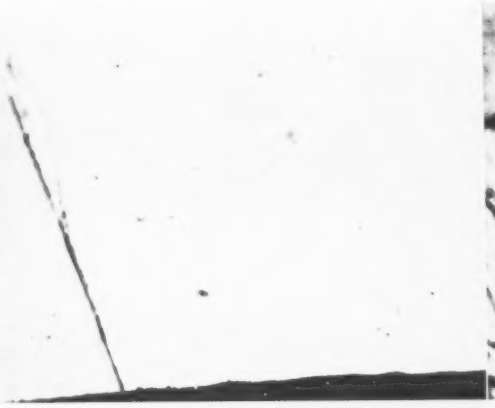


Fig. 23—Grain interface of Fe–C alloy heated to 1390° C. × 500

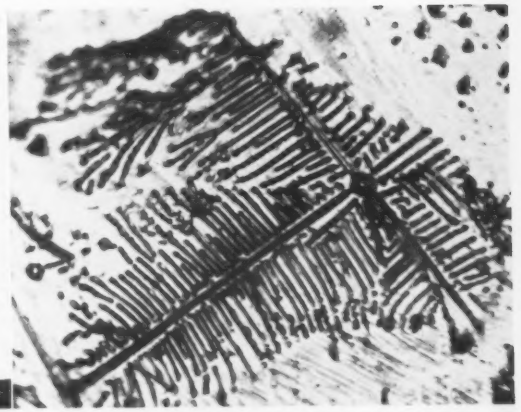


Fig. 24—Grain interface inclusions in Fe–C–Mn–S alloy heated above 1380° C. × 800

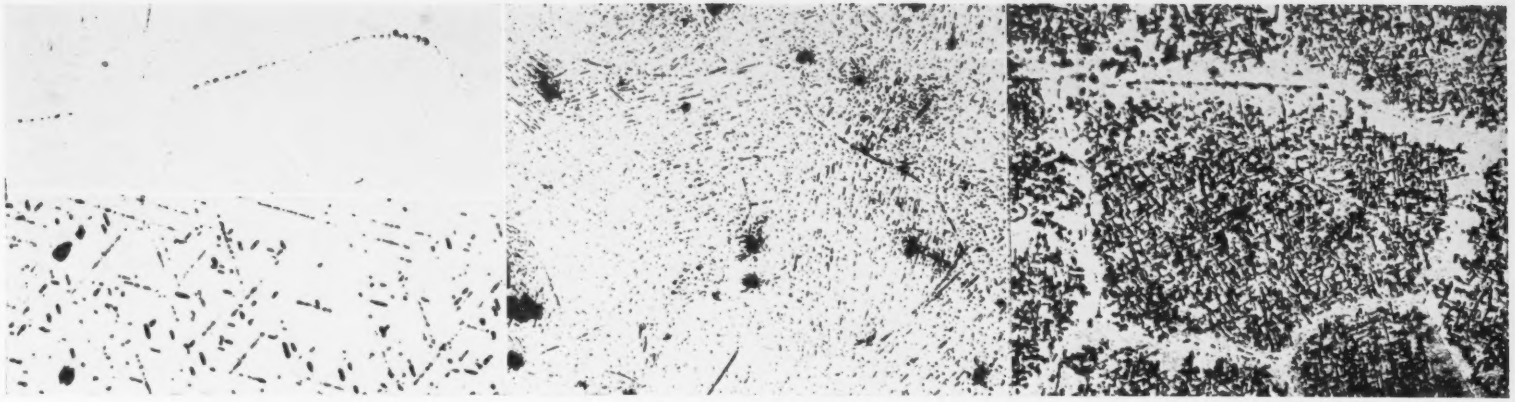


Fig. 25—Inclusions in Fe–C–Mn–S alloy (*a*) heated to 1410° C., (grain boundary inclusions), (*b*) heated to 1410° C., etched with 2% sulphuric acid ; (fine inclusions), (*c*) heated above 1350° C. Etched with nitro-sulphuric acid, (*d*) heated to 1410° C. Etched with ammonium nitrate

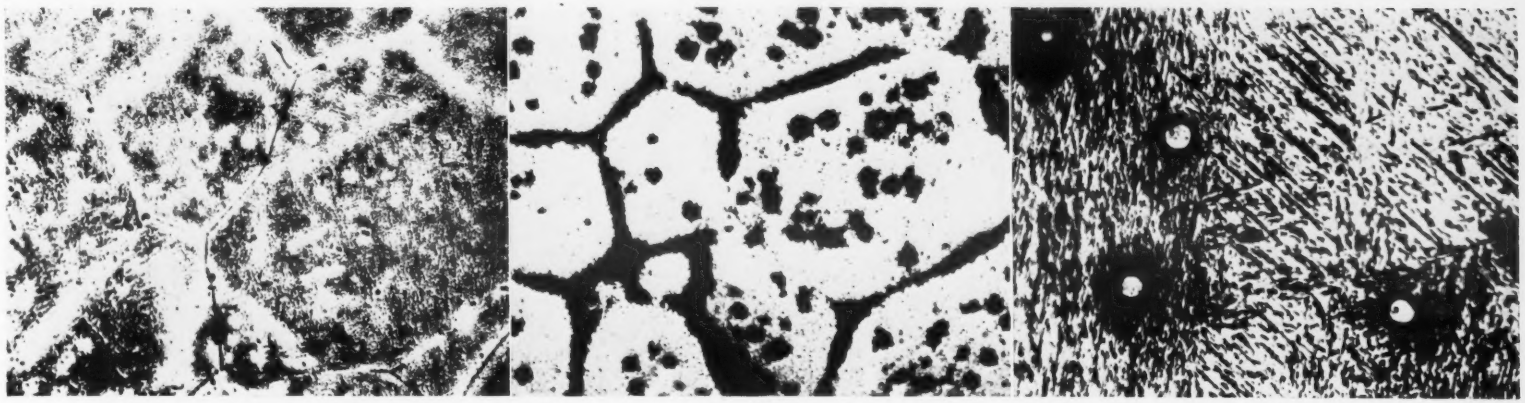


Fig. 26—Fe–C–P alloy (*a*) segregation of phosphorus in alloy heated to 1435° C., quenched and tempered. Etched with 2% nital, (*b*) heated to 1445° C. Etched with ammonium nitrate, (*c*) iron phosphide after heating to 1445° C. and heat-tinting

Fig. 27—Cleavage fracture in low carbon steel × 500

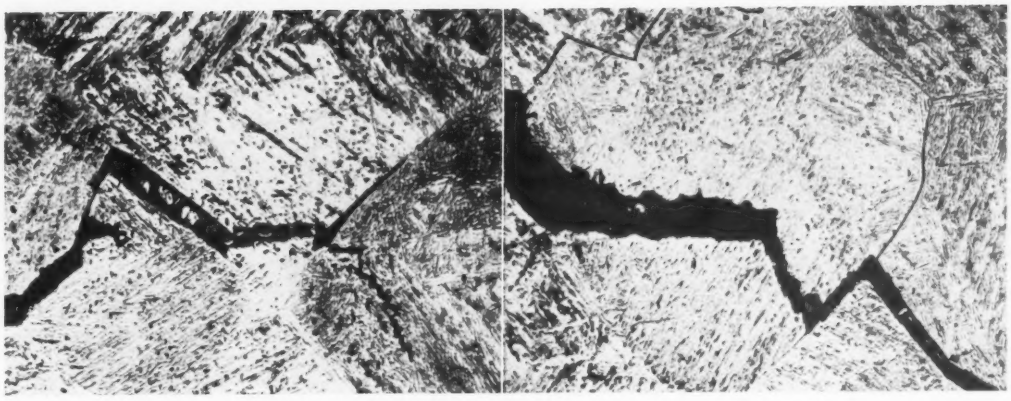
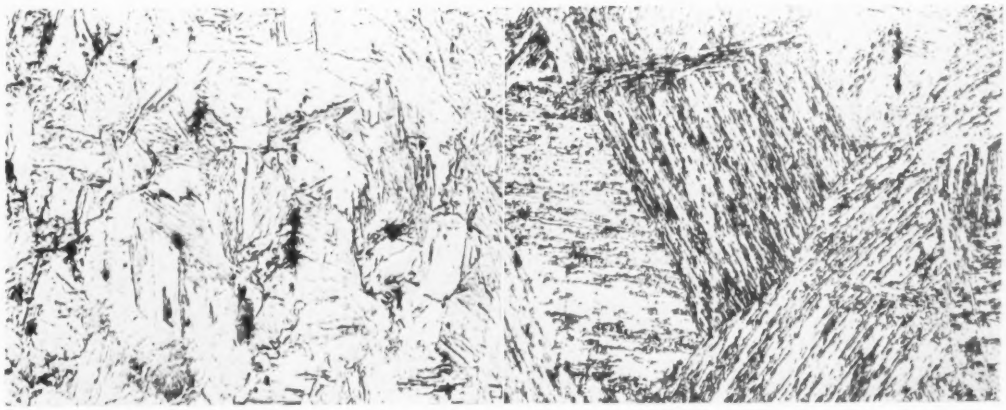


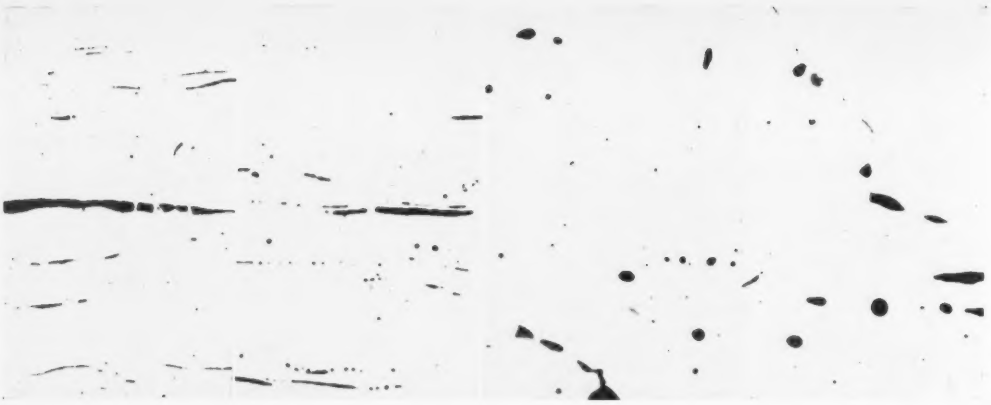
Fig. 28*a*—Crack near the fracture in specimen of steel *N*, heated to 1480° C. × 200

Fig. 28*b*—Another crack in the specimen shown in Fig. 28*a* × 150



(*a*) × 500 (*b*) × 200

Fig. 29—Steel *N* brine-quenched from 950° C. Etched with 2% nital (*a*) heated to 950° C., (*b*) heated to 1410° C.



(*a*) × 500 (*b*) × 500 (*c*) × 500

Fig. 30—Inclusions in steel *HS* heated (*a*) to 950° C., (*b*) to 1415° C., (*c*) to 1450° C.

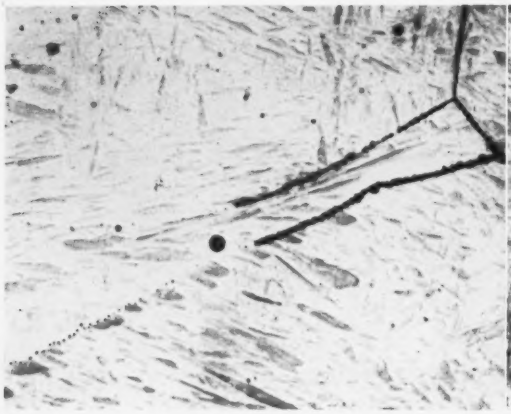


Fig. 31—Grain-boundary inclusions and intercrystalline cracks in steel *P* heated to 1430° C. × 500

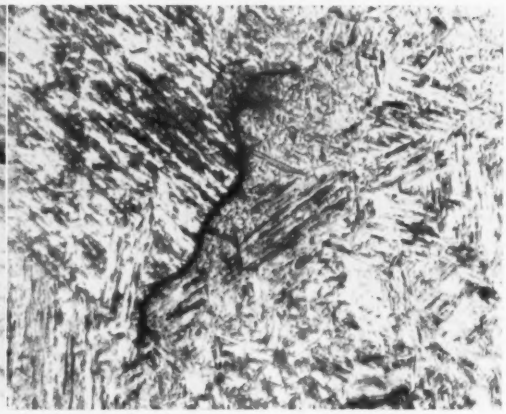


Fig. 32—Fe–C alloy quenched from 950° C. Etched with nitro-sulphuric acid × 120

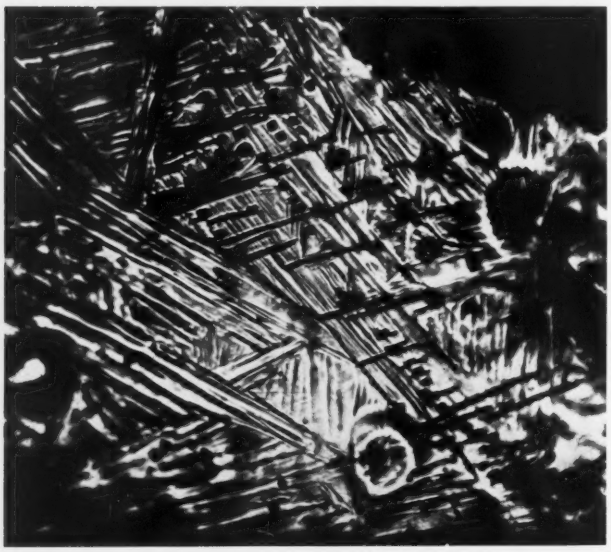


Fig. 33—Grain interface of steel *M* heated to 1490° C., quenched × 500

temperature : direct quenching from the high temperature increased the amount of round inclusions.

The inclusions on the same facet were not always attached to the same side of the fracture. Where the inclusions were present on one side of the fracture, the surface of the facet on the opposite side of the fracture showed corresponding indents. The inclusions were usually grey in colour with a yellowish tint, but the ferns sometimes showed interference colours. Chemical identification, by the method recommended by the American Society for Metals,[13] and reaction between these inclusions and dilute sulphuric acid, indicated that they were composed of a mixed sulphide of Mn and Fe, the actual composition varying from place to place. Their sections could be seen in the normal microsections of the specimens as fine inclusions, 0·0001 × 0·0005 mm. approx. in size and yellow in colour, at the austenite grain boundaries. That they were sulphide was confirmed by heat-tinting and by the use of Whiteley's tartar-emetic reagent.

In the as-quenched condition, when etched with 2% nital, the specimens showed a martensitic structure. When the specimens were tempered and then etched, regions near the grain boundaries and areas around many Fe-rich sulphide inclusions became white-etching regions ; the contrast increased with increasing tempering temperature and with increasing carbon content. The structure of the tempered martensite was continuous, extending from the dark-etching matrix into the white-etching regions. The white-etching areas were more diffuse in specimens slowly cooled from high temperatures (Fig. 14*a*), and those around inclusions were more numerous when the specimens were directly quenched (Fig. 14*b*).

No white-etching areas were found associated with the massive, dark grey sulphide inclusions. The inclusions associated with white-etching regions were light grey or yellow, and sometimes duplex with a white constituent. The light grey and the yellow constituents were found to be sulphides. The white constituent was not attacked by Whiteley's tartar-emetic reagent, and its chemical behaviour towards all reagents was identical with that of the white iron phosphide found in an Fe–C–P alloy, heated to temperatures above the solidus (described on p. 57).

That the white-etching areas were richer in a certain element or elements was shown by the fact that little copper was deposited on the areas when Stead's copper reagent (cold) was applied (Fig. 14*c*). Scratch and microhardness tests showed that the white-etching areas were harder than the matrix. This is illustrated by Fig. 15, the indentations of which were made with a Burgsmann tester and a V.D.P. indentor on a specimen of steel *VB*, air cooled from 1490° C.

A normal indication of the overheating structure (Fig. 16*a*) was obtained with nitro-sulphuric acid on specimens in the as-quenched condition. When the specimens were tempered, however, the white-etching areas remained white, with numerous pits situated along the grain boundaries (Fig. 16*b*). On the other hand, when the specimens were heat-treated and etched with ammonium nitrate, a structure consisting of a black network on a white background was obtained (Fig. 17). This reversal of the etching behaviour of specimens heated to very high temperatures has been previously observed by Preece *et al.*[9]

When a specimen of steel *T* was reheated to 900° C. and air-cooled, light etching bands were found, running across the pearlite colonies (Fig. 18) : the pearlite lamellae were continuous and no non-pearlitic structure could be found. This showed conclusively that the appearance of the white-etching areas is not due to decarburization, as has been suggested by some investigators.[5]

The conclusion that sulphide inclusions are precipitated on the austenite grain interface when a steel is overheated can be explained on the assumption that the solubility of sulphur in austenite increases with increasing temperature. Then, when a steel is cooled after being heated to high temperatures within the austenitic range, the excess sulphur will be precipitated as sulphide at the austenite grain boundaries. The amount of the grain-boundary precipitate will increase with increasing concentration of dissolved sulphur in the steel, and hence with increasing temperature. At very high temperatures, (stage 3) the solidus/liquidus range is entered, and a liquid solution of sulphur in iron is formed, which freezes as a sulphide and iron eutectic. Segregation of impurities in the liquid will occur and give a white-etching network, with nital etch, after cooling.

Effect of Slow Cooling and Long Annealing—Cooling the specimen from stage 3 at a rate of 1·5° C./min. to 950° C. before quenching reduced the inclusions on the intercrystalline surface to a few large areas of grey, transparent sulphide, as shown in Fig. 19. No white-etching network and overheated structure were observed.

When the specimens were cooled from stage 3 to stage 2 at the same rate, and then cooled rapidly in the low-temperature furnace to 950° C. before quenching, some fine inclusions were present at the grain interface together with the large ones. The nitro-sulphuric acid etch showed traces of overheating.

These observations are in agreement with Preece's findings,[9] that overheating of steel can be suppressed by slow cooling through the overheating range. Precipitation of inclusions on the grain interface still occurs, but only on a few nuclei.

Comparison of the sulphide distribution in the grain interiors of these specimens with that in the rapidly cooled specimen showed that when the specimen was cooled rapidly, the inclusions were partly segregated at the grain boundaries and partly randomly distributed in the grains ; few remnants of the inclusion strings in the grains could be seen (Fig. 20*a*), but when the specimen was cooled very slowly from the high temperature, the inclusion strings reappeared in the microsection, and the size of the individual inclusions in the strings increased with the length of the temperature range through which the steel had been slowly cooled. Thus, the inclusions within the grains of a specimen slowly cooled to 950° C. (Fig. 20*c*) were much larger than those in the specimen slowly cooled only to 1300° C. (Fig. 20*b*). This shows that, although the traces of the original

inclusions are invisible in the rapidly cooled specimen, they can still serve as active centres of precipitation when conditions are favourable.

Annealing specimens in stage 3 at a temperature just below the overheating temperature caused the coalescence of the precipitates on the grain interface to a few large inclusions, similar to those in Fig. 19. No overheated structure was observed, the inclusion strings in the grains did not reappear, and dark grey manganese sulphide inclusions were randomly distributed in the grains. The annealing also caused the replacement of Fe by Mn in the inclusions and the homogenization of the grain-boundary segregation.

Effect of γ–δ *Transformation* (*Steel M*)—The following observations were made on tensile specimens of steel *M* which had been heated at very high temperatures :

(i) When specimens were quenched from 950° C. after being heated above 1480° C., some of the inclusions became aligned in rows on the intercrystalline surface (Fig. 21)

(ii) In the microsections, the distribution of inclusions depended on the temperature range to which the steel had been heated and on the rate of cooling

(iii) Cooling to 950° C. in the low-temperature furnace from below 1470° C., before quenching, caused the inclusions to become both randomly distributed in the austenite grains and segregated along the boundaries. Direct quenching from above 1480° C., however, produced numerous inclusions in the networks running across the austenite grains (Fig. 22)

(iv) Cooling in the low-temperature furnace from above 1480° C. produced several generations of the inclusion networks, both across the austenite grains and along the grain boundaries.

These observations are best explained by the assumption that, in steel, the solubility of sulphur is lower in δ-iron than in γ-iron at a given temperature. Then, when the steel is heated into the δ range, the sulphur dissolved in austenite will precipitate preferably at the grain boundaries, where the segregation of sulphur has occurred and where the δ phase will be first formed. During cooling, part of these inclusions will first redissolve in the iron when it changes from δ to γ form and then reprecipitate along the new austenite grain boundaries on further cooling. However, a new generation of austenite has been formed by this transformation, and sheets of inclusions delineating the old boundaries will now lie across the new austenite grains, forming a transcrystalline network. The amount redissolved depends upon the rate of cooling through the austenitic range : when this is low, enough sulphide will be redissolved and subsequently reprecipitated on the new grain interface to give rise to a second generation of the inclusion network.

Iron–Carbon Alloys

In the above investigations identification of the inclusions was made by chemical methods and by heat-tinting. Attempts to identify these inclusions by the electron diffraction method were unsuccessful, owing to the roughness of the fracture. It was thought that a direct confirmation of the above conclusions could be obtained by comparison of the intercrystalline surfaces of Fe–C and Fe–C–Mn–S alloys, prepared from pure materials under identical conditions. Accordingly, alloys of various compositions were made, and their intercrystalline surfaces were examined after heat-treatment.

*Iron–Carbon Alloy F*1—Specimens of the alloy *F*1, containing only Fe and C, when heated to different temperatures between 950° C. and 1430° C., gave very coarse and clean surfaces ; nothing could be observed on the facets (Fig. 23). No white-etching network could be found with the nital etch and, when the specimens were etched with nitro-sulphuric acid, no overheated structure was obtained.

*Iron–Carbon–Manganese–Sulphur Alloy F*2—Specimens of Fe–C–Mn–S alloy, heated to temperatures up to 1300° C., had very clean intercrystalline surfaces, with only a few fine inclusions. Many dark grey inclusions of irregular shapes, soluble in dilute sulphuric acid and in boiling alkaline sodium picrate, were found on the fracture surface of a specimen heated to 1350° C. ; however, no inclusion could be seen along the grain boundaries in a polished section at 1000 magnification.

The fractures obtained in specimens heated above 1380° C. contained inclusions of both particle and fern types (Fig. 24), the latter increasing with increasing temperature. The shape and the colour of these inclusions were identical with those observed in plain carbon steels, and chemical identification showed that they were mixed sulphides of Mn and Fe. Sections of these fern inclusions could be seen in the microstructure as fine sulphide inclusions, yellow in colour, lying along the grain boundaries (Fig. 25*a*).

A new type of precipitation of inclusions on the crystallographic planes of the austenite was revealed in the specimens heated above 1350° C., the amount increasing with increasing temperature. The distribution of these inclusions is clearly shown in Fig. 25*b* ; the specimen had been etched with 2% sulphuric acid and repolished, the traces in the photograph consisting of enlarged pits after the removal of inclusions. These inclusions, of non-uniform width, were not clearly seen under an oil-immersion objective without light etching, *e.g.*, with picral, but appeared yellowish-grey at the wider portions. They could not be removed by reheating the specimens to 950° C. for 30 min. Fewer of them were found in regions near the austenite grain boundaries, where fine sulphide inclusions were present, than in more remote regions. From their response to various chemical reagents and their relation to the grain-boundary inclusions, it is believed that these precipitates are sulphide, similar to the fine sulphide dispersions observed by Preece and Irvine,[15] in their investigation of sulphide penetration in steel. The presence of these fine sulphide inclusions in a Widmanstätten pattern indicates that they were formed by precipitation. That the orientations of these inclusions are constant in one austenite grain but different in different grains suggests that they were precipitated during cooling. Had they been formed during heating or at high temperatures, the migration of the grain boundaries due to grain growth after the precipitation would not give this consistency of orientation in a grain. Some sulphur must, therefore, have entered into solution at 1350° C. in the alloy *F*2. These fine precipitates were also present in specimens of steels *P*, *N*, and *M* heated above 1400° C.

Traces of the overheating structure were obtained, when etched with nitro-sulphuric acid, in specimens heated above 1350° C. (Fig. 25*c*), the intensity of overheating increasing with increasing temperature. A normal martensitic structure was obtained in all specimens in the as-quenched condition with the nital etch, but when the specimens were tempered, those heated to 1380° and 1400° C. showed very faint white-etching areas around some light-coloured sulphide inclusions and along a few grain boundaries. In the specimen heated to 1440° C., however, in spite of the presence of numerous fine light-coloured sulphide inclusions along the grain boundaries, no white-etching network was observed. Reversal of the structure, as found in steels heated to stage 3 when etched with ammonium nitrate, was not observed in *these* specimens. The structure obtained consisted of a network, immune to etching, in a black background outlined by the grain-boundary inclusions (Fig. 25*d*).

*Iron–Carbon–Phosphorus Alloy F*3—The inconsistent results obtained in the Fe–C–Mn–S alloy, in connection with the appearance of the white-etching regions when etched with nital, suggested that their appearance was due to some secondary effect, such as the presence of some particular impurity.

One of the elements almost absent in the alloys *F*1 and *F*2 but present in all steels is phosphorus. This is consistent with Stead's[2] suggestion that the appearance of the white-etching network in a burnt steel may be due to the segregation of phosphorus. Accordingly, an Fe–C–P alloy was prepared and studied. The alloy, forged after annealing for 2 hr. at 1000° C., showed coring of phosphorus when quenched and tempered, but was completely homogenized after heating for 15 min. at 1380° C.

The specimens, heated to different temperatures and quenched from 950° C., gave very clean fracture surfaces. Microstructures of all specimens heated to temperatures up to 1420° C., when etched with 2% nital, were martensitic and showed no white-etching network when tempered.

A few white-etching areas and one faint white-etching boundary appeared in the specimen heated to 1425° C. A white-etching network nearly identical with that obtained in the steels, and numerous white-etching areas in the grains, were found in specimens heated above 1425° C. (Fig. 26*a*). Intergranular melting became evident at 1445° C., but there was no difference in the general structure. No fine inclusions could be found at the austenite grain boundaries.

When the specimens were etched with nitro-sulphuric acid the white-etching areas remained white, in a black matrix, and cracks opened up, giving a structure similar to that obtained in a steel specimen in stage 3. However, unlike the latter (Fig. 16*b*), there were no black pits within the white-etching boundary network, and the grain boundaries which were not cracked remained intact. The ammonium nitrate etch gave a black network on a white background, Fig. 26*b*.

The white-etching network obtained with nital etch in this alloy differed from that in plain carbon steels which always followed the austenite grain boundaries, in that it sometimes deviated from them. This is understandable when the effect of the grain-boundary inclusions in hindering grain growth is considered. The grain boundaries in this alloy, being free from inclusions, could move more freely during cooling, away from the original position existing at the high temperature where segregation had occurred.

The round inclusions situated within the white-etching areas, Fig. 26*c*, were not attacked by dilute organic acids or by sulphuric acid ; they remained white on heat-tinting up to about 450° C., above which they became slightly yellowish ; they were stained when etched in boiling neutral sodium picrate for one hour, and were removed by boiling alkaline sodium picrate. All this evidence indicates that they were iron phosphide.

The solidus of an Fe–C alloy of the same carbon content as this alloy, according to Adcock,[16] is 1427° C. approx. The specimens in which the white-etching network and areas were found, therefore, had all been heated to temperatures above the solidus. When the solidus/liquidus range is entered, a P-rich liquid phase will be formed at the grain boundaries and at isolated points in the grains, and when the alloy is subsequently cooled at a rate sufficiently rapid to prevent complete homogenization P-rich iron will freeze, giving rise to the P-rich white-etching network and regions. The phosphorus concentration in the remaining liquid will be increased by the precipitation of the solid solution, and when the eutectic temperature is reached iron phosphide will finally separate and exist as inclusions in the alloy. No iron phosphide was observed at the grain interface ; the large contact area favours the diffusion of phosphorus, which must be very small in amount, and, presumably, the iron phosphide is removed by the partial homogenization resulting from such diffusion.

These experiments on Fe–C alloys show that the formation of the light and dark etching networks, with nital and ammonium nitrate respectively, is due to the segregation of phosphorus. Other elements may behave in a similar manner, but whether a boundary network will remain after cooling depends upon the rates of diffusion of the elements concerned, and on the rate of cooling from high temperatures. The occurrence of the white-etching network revealed by nital etch in steel specimens heated to stage 3 is, therefore, not directly associated with overheating, but is due to the formation in the steel of a phase in which phosphorus has a higher solubility than in the solid iron.

DISCUSSION

Grain-Boundary Precipitation and Overheating

The results of previous and present researches into the effect of various factors on the overheating of steel, the observation in the present investigation of the precipitation of sulphide inclusions at the austenite grain interface in steels and in an Fe–C–Mn–S alloy, the absence of such precipitation and overheating effect in Fe–C and Fe–C–P alloys, and the fact

that increasing the cooling rate decreases the amount of precipitation in the grain interface, all show that overheating is a precipitation phenomenon (as was first postulated by Woolman and Kirkby[10]) involving principally sulphide.

The solubility of sulphur in austenite is not known. The authors have observed changes in the number and distribution of sulphide inclusions at high temperatures which indicate that sulphur is soluble in austenite to some extent. Assuming that the solubility of sulphur in austenite increases with increasing temperature in the overheating range, when a steel is cooled after being heated to a temperature in this range, excess sulphur will be precipitated on any existing nuclei, or at places such as grain boundaries where nucleation is most likely to occur. The amount of inclusions precipitated at the grain boundaries will increase with increasing temperature, because (*a*) the amount of sulphur in solution, which precipitates on subsequent cooling, is raised, and (*b*) more sulphide particles are dissolved and there is less grain-boundary area (due to the increased austenite grain size), so that fewer sites are available for precipitation on cooling, and, hence, the precipitation on these sites (*i.e.*, grain boundaries and undissolved particles) will be more intense. It is the consequence of this mechanism that overheating is usually less severe in steels of high sulphur content,[9] because more sulphide particles remain undissolved in these steels, thus providing centres of precipitation other than the grain boundaries.

The intensity of the sulphide precipitation at the austenite grain interface, which controls the degree of overheating, will also depend directly on many other factors, *e.g.*, the rate of cooling, the solubility and diffusivity of sulphur in the austenite, and hence on the composition of the steel, and, indirectly, on the de-oxidation, casting, soaking, and rolling practices.

The precipitation of a large number of fine sulphide inclusions at the grain boundaries will lower the breaking strength, and, therefore, the ductility and toughness of the steel. The inclusions precipitated on the grain interface at relatively high temperatures cannot be removed or substantially altered by heating to a lower temperature for short periods ; thus, when the overheated steel is suitably heat-treated (to give a tough matrix), the steel will subsequently rupture on the surfaces on which precipitation of inclusions has occurred during overheating, giving rise to the familiar faceted fracture. The presence of many acid-soluble inclusions will cause the formation of the cellular structure when etched with nitro-sulphuric acid. The electrochemical behaviour of the matrix in the immediate neighbourhood of these inclusions must be affected by their presence, and this may give rise to the etching effects observed with other specific chemical reagents, *e.g.*, ammonium nitrate and molten zinc chloride as used by Preece *et al.*[9] The overheating temperature of a steel, determined by etching, will depend on the sensitivity of the method used in its determination, since precipitation of inclusions at the grain interface occurs before the overheated structure can be revealed in a polished section.

Formation of a Eutectic and the Segregation of Phosphorus

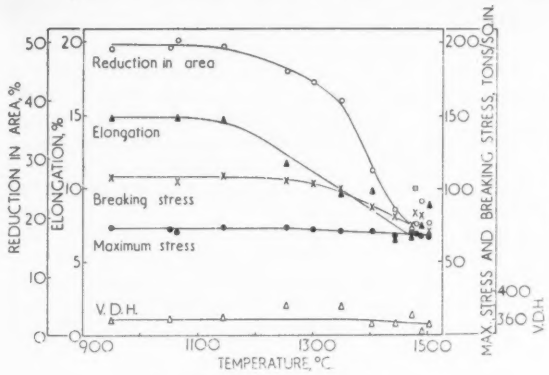
The observation that, in steel heated to stage 3, the sulphide is found at the grain interface as a component of a eutectic suggests that liquid was present at high temperatures at these places, *i.e.*, melting had taken place in the regions of the grain boundaries. The lowest austenitizing temperatures that produced specimens in which either the fern inclusions or the white-etching network, with nital etch, or both, were found, are given below, together with the solidus of Fe–C alloys (containing about 0·0027% of S) of the same carbon contents.[16]

Steel	Carbon, %	Solidus, °C.	Lowest temp., °C.	Difference, °C.	Overheating temp., approx., °C.
T	0·85	1380	1350	30	1300
S	0·47	1440	1420	20	1350
P	0·23	1480	1430	50	1360
N	0·17	1490	1470	20	1390
Fe–C–Mn–S	0·48	1440	1380	60	1350
Fe–C–P	0·56	1427	1425		

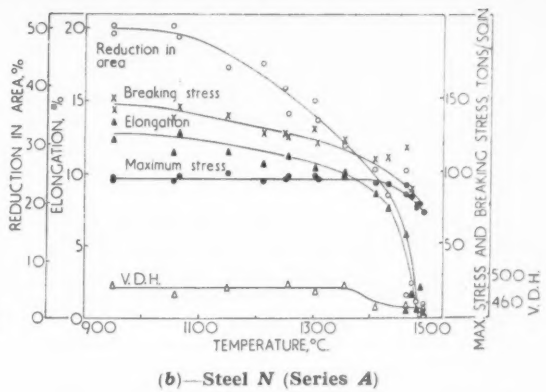
The temperatures at which the eutectic inclusions appeared in the sulphur-bearing materials are well below the solidus of Fe–C alloys of the same composition but with little sulphur. Indeed, the melting of iron at the grain boundaries at a temperature 60° C. below the solidus is difficult to conceive. If steel at high temperatures contains a certain amount of sulphur in solution, then its solidus will be lower than that of an Fe–C alloy free from sulphur. The solidus line of the Fe–C–Mn–S alloy is not known, but it must be very steep, because (*a*) sulphur forms eutectics of low melting points (*e.g.*, 1000° C. for the Fe–FeS eutectic and 1180° C. for the FeS–MnS eutectic,[17] and (*b*) the maximum solubility of sulphur in solid iron must be very small. Hence, it is possible to cross the solidus of a steel, containing a small amount of sulphur in solution, at a temperature well below the solidus for a pure Fe–C alloy. It is suggested, therefore, that stage 3 is characterized by the entry of the steel into the solidus/liquidus range, giving the formation of a small amount of liquid solution rich in sulphur. This liquid begins to form at places where the sulphur content is high, such as grain boundaries and regions near the sulphide inclusions.

When the steel is subsequently cooled at a fairly rapid rate, the sulphur partly redissolves in the solid, but the bulk remains at the grain interface, and eventually separates as an Fe–FeS eutectic when the eutectic temperature is reached. Subsequent conversion during cooling may change some of the Fe-rich sulphide into manganese sulphide. The eutectic structure obtained will depend on the amount of other elements, such as carbon, dissolved in the liquid, and on the rate of cooling. If the steel is cooled extremely slowly through the high-temperature range equilibrium may be approached or established, with the result that only solid Mn-rich sulphide will precipitate, and no fern inclusions will form.

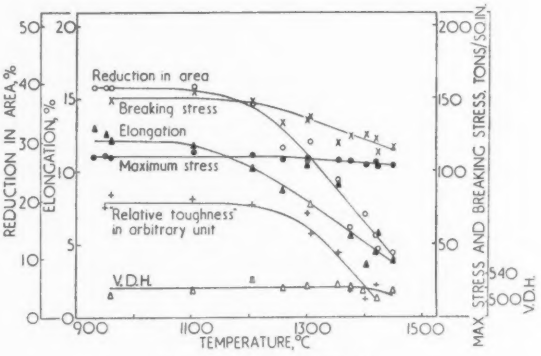
When a liquid Fe–S solution is present along the grain boundaries or in the grains, segregation of phosphorus and other elements into the liquid will occur. Phosphorus can segregate particularly easily in this manner because it is much more soluble in liquid iron than in solid iron. On cooling, P-rich iron will



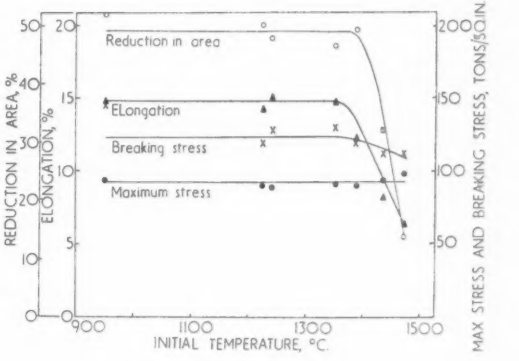
(*a*)—Steel ***M*** **(Series *A*)**



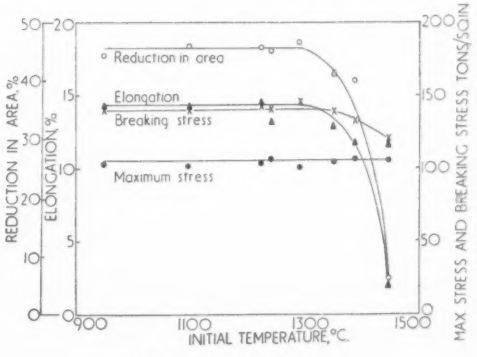
(*b*)—Steel ***N*** **(Series *A*)**



(*c*)—Steel ***P*** **(Series *A*)**



(*d*)—Steel ***N*** **(Series *B*)**



(*e*)—Steel ***P*** **(Series *B*)**

Fig. 34—Effect of high-temperature treatment on mechanical properties of different steels: (*a*)–(*c*) series *A*, quenched from 950° C. without normalizing treatment ; (*d*), (*e*) series *B*, quenched from 950° C. after normalizing treatment.

be deposited around the inclusions and at the grain boundaries, and, after quenching, a P-rich iron network will be present, which, being more resistant to tempering, will appear as a light-etching network with the nital etch. The ' width ' and the diffuseness of the network will depend upon the degree of homogenization which takes place after segregation, and increases with decreasing cooling rate.

The above explanation of the phenomena observed when the steel has been heated to stage 3 is a development of that usually given for burning, and is similar to that used by Whiteley[14] to explain the conversion of MnS to FeS when a steel is heated to temperatures above the solidus. Steels which have been heated to stage 3, with the formation of a eutectic at the grain interface and of a white-etching network revealed with the nital etch, can, therefore, be adequately described as ' burnt,' in the sense that a liquid has been formed during heating.

MECHANICAL PROPERTIES OF MARTENSITIC LOW-CARBON STEEL

The effect of overheating on the mechanical properties of steel has been investigated by many investigators, among whom, Austin[7] has shown that the low ductility of an overheated steel in a tensile test is a result of the premature termination of load/elongation curves by fracture. The effect of high-temperature treatment below the overheating temperature on the mechanical properties is less known ; systematic investigations are therefore desirable, if the effects of high-temperature treatment are to be fully understood.

The tensile test was used in the present investigation ; the low hardenability of these steels requires the use of small specimens with diameters below $\frac{1}{8}$ in. and renders any other study very difficult. Two series of tests were made ; series *A*, covering all three low-carbon steels, *P*, *N*, and *M*, was obtained by quenching the specimens from 950° C., after the specimens

had been cooled from high temperature to 950° C. in the low-temperature furnace. In series *B*, which included two plain carbon steels, *P* and *N*, and one free-cutting steel, *HS*, the specimens were normalized after high-temperature heating, reheated to 950° C., and quenched.

Plain Carbon Steels

The results of the tests for three carbon steels are given in Figs. 34*a*, *b*, and *c* for series *A*, and in Figs. 34*d* and *e* for series *B*, in which the properties are plotted against the highest temperatures to which the specimens had been heated.

Maximum Stress—The steels containing 0·23, 0·17, and 0·08% of carbon, when fully hardened from 950° C., have maximum stresses of 110, 95, and 73 tons/sq. in., respectively. The maximum stresses are not greatly affected by the austenitizing temperature, being slightly lowered in series *A* after heating to high temperatures. The effect is completely removed when the steel is quenched after being reheated to a constant austenitizing temperature. The small difference in the maximum stresses in the two series in steel *P* was found to be caused by a small variation of about 0·02% in the carbon content of the bars from which these specimens were machined. The carbon content reported in Table I corresponds to that of the specimens in series *A*.

Hardness—The V.D.P. hardnesses of steels *P*, *N*, and *M* at 30-kg. load are 520, 480, and 360 respectively, and closely reflect the maximum stresses of the steels.

Limit of Proportionality—The limits of proportionality of steels *P*, *N*, and *M* are 92, 76, and 59 tons/sq. in., respectively, and show no systematic variation with austenitizing temperature.

Breaking Stress—The breaking stress, calculated from the actual load at rupture and the cross-sectional area of the ruptured specimen, decreases with increasing temperature in series *A*. In series *B*, the breaking stress remains practically constant until the steel is overheated.

Reduction in Area and Elongation—The reduction in area and the elongation (on 0·447 in.) of the specimens in series *A* decrease slowly at first, and then rapidly with increasing temperature; the steels *P* and *N* become virtually brittle after being heated above 1450° C. The effect can be completely removed by a single normalizing treatment when the previous heating has been below the overheating temperature, but becomes persistent when the steels are overheated. The curve marked 'relative toughness' in Fig. 34*c* shows the effect of high-temperature treatment on the area under the load/elongation curve obtained during the testing.

Fracture—A cup-and-cone fracture occurred in all specimens possessing considerable ductility. This changed to a flat fracture when the reduction in area became below about 18%, consisting of dull and shiny parts. Examination at high magnification showed that the bright spots were regions of cleavage fracture, as shown in Fig. 27, taken from a specimen of steel *M* in series *A*. Intercrystalline fractures were common when the steel in series *A* was severely overheated.

The juxtaposition of intercrystalline and transcrystalline fractures is shown in Fig. 28, taken from a specimen of steel *N* in series *A* heated to 1480° C. These photomicrographs clearly show that a crack may change during propagation from an intercrystalline to a transcrystalline path, and *vice versa*, or may even diverge into intercrystalline and transcrystalline cracks. The actual manner in which the steel at a particular place will rupture depends upon the relative resistance to fracture of the grain boundary and of the cleavage plane of the martensite, and upon the resolved stresses normal to these surfaces. A crack initiated at a grain boundary may result in a transcrystalline fracture when conditions are favourable.

Microstructure—All the specimens as-quenched showed the usual martensitic structure with 2% nital etchant. Figures 29*a* and *b* show the microstructures of steel *N* in series *A*, heated to 950° C. and 1410° C. respectively. It will be seen that, when the steel had been heated to high temperatures, the martensite became more regularly arranged within the austenite grains. This regularity of the martensite can be removed by reheating the steel into the austenitic range and then quenching. The martensite obtained in steel *M* by directly quenching the specimen from the δ range, corresponds to that obtained from a fine-grained austenite.

The overheated structure was obtained in all those specimens of series *B* which showed a loss of ductility, and in those of series *A* which had been treated above the temperatures at which an overheated structure was found in series *B*. A faint white-etching network was observed, after etching with nital, in the tempered specimens of steels *P* and *N* which had been heated to temperatures above 1430° and 1470° C., respectively. The appearance of the white-etching network was accompanied by a reversal of the overheated structure to a white network on a dark background, when the specimen was etched with nitro-sulphuric acid.

Free-Cutting Steel

The idea that grain-boundary precipitation of sulphide is responsible for overheating led to an investigation of the effect of high-temperature treatment on the properties of free-cutting steel, in which sulphide inclusions are abundant.

Tensile specimens of a commercial free-cutting steel, *HS*, were heat-treated as in series *B* of the plain carbon steels. The tensile properties were affected when the steel had been heated above 1400° C., and above 1450° C. the steel became practically brittle. Considerable change in the size and the distribution of the sulphide occurred in the specimens heated above 1415° C., as can be seen from Figs. 30*a*, *b*, and *c* taken from sections parallel to the rolling direction. The original sulphide inclusions in the steel were massive, were elongated in the rolling direction, and were sometimes broken up into fragments: very few fine inclusions could be seen. In the specimens heated to 1415° and 1420° C., however, numerous fine inclusions, in strings, could be found. These inclusions are thought to be the remnants of the massive inclusions after solution of sulphide

in the austenite. At still higher temperatures, some of the inclusions became globules, as shown in Fig. 30*c*.

The nitro-sulphuric acid etch revealed the overheated structure in all specimens heated above 1415° C., but the ammonium nitrate etch gave a structure only in the specimen heated to 1450° C., in which a black network corresponding to that obtained in steels in stage 3 was observed.

Iron–Carbon–Manganese Alloy

The results obtained above showed that the effect of increasing austenitizing temperature on the ductility of the steels can be divided into two stages. The decrease of ductility above the overheating temperature is persistent in respect of a single normalizing treatment, and must be caused by precipitation of sulphide inclusions at the grain interface, which undoubtedly weakens the intercrystalline cohesion and initiates cracks. The weakening effect of the grain-boundary inclusions is shown by the association of cracks with inclusions (Fig. 31). The precipitation of fine inclusions had stopped near a large inclusion, and the propagation of the lower crack was interrupted at the same place.

The embrittlement of steel below the overheating temperature in series *A* required further study. The loss of ductility cannot be due to the presence of micro- or macro-stresses, because the steel, as-quenched, possesses considerable ductility, and any difference in such stresses should be mostly, if not completely, eliminated by the plastic deformation in the initial stages of the tensile test. Furthermore, specimens tempered at 400° C. for 1 hr. after the high-temperature treatment showed similar deterioration in ductility. That the low ductility of the specimens was not caused by the presence of quench-cracks in the coarse-grained martensite was shown by the fact that the loss of ductility in a specimen quenched after high-temperature heating (below the overheating range) could be recovered by reheating and quenching from 950° C.

The loss of ductility, therefore, is caused either by the precipitation of fine inclusions in too small amount to produce a visible overheating effect, or by the large austenite grain size. Preliminary investigation of the grain growth of steel *N*, under the same conditions as in series *A*, showed that the grain size after heating for 60 min. at 1200° C. was the same as that after only 3 min. at 1250° C. The specimens heated for increasing periods up to 60 min. at 1200° C. showed a decrease in the reduction in area, from 42 to 35%, but the difference in the elongation was too small to be conclusive.

It was thought that an Fe–C alloy containing no sulphide, which should be free from grain-boundary precipitation, could be used to determine whether the loss of ductility in steel is due to this precipitation process. An Fe–C alloy, containing 0·24% of C (*F*4) was prepared, and tensile specimens cut from forged bars were heat-treated. All specimens of this alloy, quenched from 950° C., ruptured within the elastic limit at about 60 tons/sq. in. Both intercrystalline and transcrystalline surfaces were present in the fracture, the former being very clean and highly reflective. The specimens showed a normal martensite structure when etched with 2% nital, but gave, with nitro-sulphuric acid, several continuous lines of lengths between 0·005 and 0·01 in. (Fig. 32).

An Fe–C–Mn alloy (*F*5) was next prepared, and tensile specimens, machined from homogenized and forged bars, were heat-treated in two series, as in the case of plain carbon steel. The results are given below :

Temp., ° C.	Austenitic Grain Size, A.S.T.M. No.	Maximum Stress, tons/sq. in.	Elongation, % on 0·447 in.	Reduction in Area, %
Series *A*				
950	3	98	9	30
1290	2	93	10	30
1350	−1	96	2	10
1400	−2	87	1	2
Series *B*				
950	3	90	9	31
1350	3	95	9	31
1440	3	97	8	36
1480	3	94	9	32

The last two specimens in series *A* ruptured with brittle, coarse, transcrystalline fractures, while the rest broke with cup-and-cone fractures. It will be seen that there is no loss of ductility in series *B*, or in other words, the alloy is immune from overheating : this was confirmed by etching with nitro-sulphuric acid. Nevertheless, the ductility of the specimens in series *A* decreases with increasing temperatures. These experiments indicate that effects related to large austenite grain size are probably responsible for the deterioration in ductility of steel below the overheating temperature.

DISCUSSION

The hardness values of the tensile specimens of the three low-carbon steels, when quenched, agree with those reported in the literature for steels of comparable carbon contents, using very thin specimens and high cooling rates. Furthermore, the presence of any ferrite at the austenite grain boundaries always gives a specimen of lower strength and hardness. These show that full hardening of the steels has been achieved, the maximum stresses of martensitic plain carbon steels containing 0·08, 0·17, and 0·23% of carbon being 73, 95, and 110 tons/sq. in. respectively. These martensitic steels are ductile when the austenitizing temperature is low, with reductions in area of about 40–50% and elongations of 15% on 0·447 in. The ductility decreases with increasing austenitizing temperature ; the loss can be recovered by normalizing, or by requenching the steel, unless the overheating temperature has been exceeded during first heating.

The reduction in ductility when the steel has been heated above the overheating temperature is believed to be caused by precipitation at high temperatures of large numbers of sulphide inclusions at the austenite grain interface, as suggested on p. 58. The presence of these inclusions on planes of small curvature will cause stress concentration and will initiate fracture, thus promoting premature failure.

The loss of ductility in steel *M* ceases in the δ range. The precipitation of sulphide during the γ–δ transformation sets a limit to the solubility of sulphur in austenite, and hence to the amount of inclusions

precipitated during cooling, *i.e.*, to the severity of overheating.

The embrittlement of the martensitic steels below the overheating temperature is probably due to austenitic grain growth. It has been observed that the martensitic structure sometimes appears in relief on the exposed grain interface, as shown in Fig. 33 ; the directions of the martensite plates are different in different grains and on the corresponding facets of a fracture. Before fracture occurs along such an interface, the boundary regions must be either in a state of severe elastic strain or must possess innumerable small misfits or potential cracks. In either case, the austenite grain boundary in a martensitic steel will be a source of weakness, and will tend to initiate cracking when an external load is applied. In a coarse-grained martensite, both the smaller curvature of an unstable, severely strained region, and the regular arrangement of the martensite (*cf.* Fig. 29*b*) will favour the propagation of a crack once initiated : thus, a transcrystalline crack arriving at a grain boundary or an intercrystalline crack arriving at a grain-boundary junction has to change its direction before it can proceed further, and the frequency of such a change will be smaller in a coarse-grained martensite.

CONCLUSIONS

Fine-grained, martensitic, plain carbon steels containing 0·08, 0·17, and 0·23% of carbon have maximum stresses of 73, 95, and 110 tons/sq. in. respectively, and reductions in area of about 40%.

The ductility of martensitic steels decreases with increasing austenitizing temperature. The loss of ductility at the initial stage is due to increasing austenite grain size, and is recoverable. However, when the steel is overheated, the loss becomes persistent in respect of a single normalizing treatment.

Precipitation of sulphide inclusions occurs at the austenite grain boundary when a steel is overheated : the amount of inclusions so precipitated increases with increasing temperature.

When a steel has been heated to temperatures near the solidus of an Fe–C alloy of equivalent composition but containing no sulphur, Fe-rich sulphide inclusions are found, after cooling, at the grain boundaries as one constituent of a eutectic.

Segregation of phosphorus to the grain boundaries occurs when Fe-rich sulphide inclusions are present at the grain boundaries with a eutectic structure, or when a sulphur-free Fe–C–P alloy has been heated to temperatures above the solidus.

Slow cooling through the overheating range or long annealing at temperatures just below the overheating temperature reduces the number of inclusions at the grain boundaries, and removes the overheating effect in the steel.

On the basis of these observations, a theory is put forward that overheating is due to the precipitation of sulphide inclusions at the austenite grain boundaries existing at the temperature to which the steel has been heated.

It is suggested that :

(i) The solubility of sulphur in austenite increases with increasing temperature. When a steel is cooled from high temperatures at a suitable rate, the excess sulphur will precipitate as sulphide at the grain boundaries and on the residual sulphide inclusions in the steel

(ii) The severity of overheating at a certain temperature will increase with increasing amounts of the sulphide inclusions precipitated on the grain interface, which depend upon the rate of cooling and the number of residual sulphide inclusions in the grains

(iii) When a steel is heated to temperatures within the solidus/liquidus range (the solidus being lowered by the presence of sulphur in solution), a liquid Fe–S solution is formed at the grain boundaries, which will freeze as a eutectic on rapid cooling

(iv) Phosphorus segregation occurs when a liquid phase is present in the steel and produces a network of iron rich in phosphorus, after cooling. This network of P-rich material will appear as a white-etching network when the steel is tempered and etched with nital.

Acknowledgments

The authors wish to express their indebtedness to the British Iron and Steel Research Association and to the British Welding Research Association for supplying the grants to carry out this investigation ; to Professor A. Preece for valuable suggestions, and to Dr. A. H. Cottrell and Dr. J. A. Wheeler for their advice and many helpful discussions.

Thanks are due to the English Steel Corporation, Ltd., and The United Steel Companies' Central Research Department for supplying the steels ; to Mr. H. W. Hignett of The Mond Nickel Co. Ltd., for arranging the cold drawing of steels ; to Messrs. B. Jones and J. H. Slaggs of the English Steel Corporation, Ltd., for the use of microhardness testing equipment ; to Mr. E. Marks of Messrs. Richard Thomas and Baldwins, Ltd., who arranged for the aluminium and the gas analyses ; and to Messrs. P. Wingrove and A. J. Hawkes of the Metallurgy Department of the Birmingham University for the chemical analyses.

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A Wire-Drawing-Die Profilometer

By R. M. J. Withers

SYNOPSIS

The paper describes the development of an instrument designed to measure and reproduce the profiles of the bore of wire-drawing dies. It is intended for use with die bores with throat diameters in the range of 0·02 to 0·25 in. It can also be used to measure accurately the bore diameter. The limit of measurement is 0·0001 in.

Introduction

An investigation is now being made by the Wire Drawing Section of B.I.S.R.A. into the phenomenon of die wear, and the effect of die-bore profile and other parameters, such as characteristics of the material of the wire, on the wire-drawing process. In this connection, an instrument was required to measure accurately the diameter and profile of the bore, and to detect and measure deformities which arise, such as the growth of a ring cavity in the die bore. Figure 1 shows a section through a typical wire-drawing die.

The instrument described was developed to fulfil these requirements. It is designed primarily as a research tool, but the measuring technique adopted could be further developed to satisfy the wider requirement which exists in the industry for a works inspection tool. Such an instrument is needed in wire-drawing shops, for rapid inspection of bore profiles to show, for example, deformities from a given profile specification.

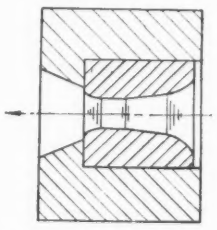

Fig. 1—Typical profile of wire-drawing die

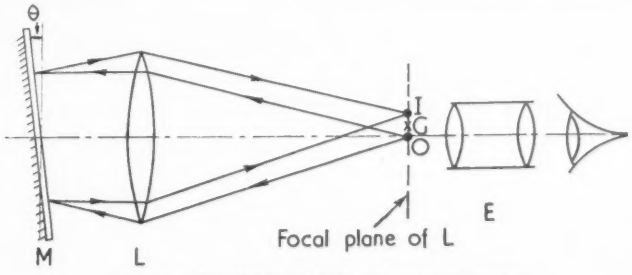

Fig. 2—Simplified sketch of autocollimator

The instrument has been designed to meet the following specification:

Range of bore diameters to be measured	= 0·02–0·25 in.
Precision of measurement	= ±0·0001 in.
Maximum amplitude of profile	= 0·25 in.
Maximum length of profile	= 1·00 in.

PRINCIPLE OF OPERATION

The system most capable of giving reliable quantitative measurements on internal bores of small diameter seemed to be that of a mechanical feeler arm, inserted into the die bore, and connected to some suitable indicating device. In its simplest form the feeler arm consists of a lever, one end of which rests on the profile while the other end moves across a scale. Linear translation of the die with respect to the lever pivot will reproduce on the scale the changes in profile. Such a technique has already been used in Germany by Mucke[1] for the measurement of dies of diameter exceeding 0·008 in., and in which the feeler arm was part of a simple lever giving a magnification of approximately 20.

The only alternative technique which has been used is that of examining, by optical projection, the shape of a partly drawn wire, or of an alloy, *e.g.*, Wood's metal, cast into the die bore, and projecting the shadow of the plug thus obtained on to a screen. Recent tests[2] have shown that the dental compound Zelex is very suitable for this purpose. Tests with this compound showed that the cone angle of the bore could be estimated to 5% and the diameter to within 1%. It seems unlikely that this technique would be suitable for determinations of greater precision.

Various types of indicating device, both electrical and optical, were considered for use with the mechanical feeler-arm technique. One of the most important limitations is the small diameter of the probe, consequent on the small diameter of the die bores: the measuring device must not produce a sufficiently large bending on the thin probe to produce a systematic error.

Of the electrical methods for detecting and indicating displacements of the feeler arm, use of a parallel

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Mr. Withers is on the staff of the British Iron and Steel Research Association.

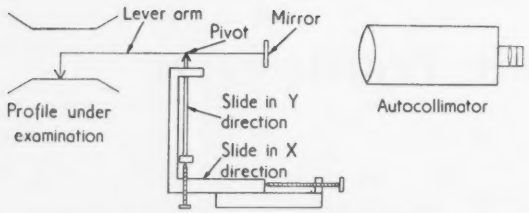

Fig. 3—Principle of measuring instrument

plate condenser was first considered. This device gives ample sensitivity, but, in order to achieve sufficient range of movement with a linear response and also to make the system independent of varying parameters of the auxiliary electronic equipment, a considerable amount of development work was expected in employing this technique.

An alternative method suggested was the use of an autocollimator, to measure the angle of rotation of the probe lever with the aid of a reflecting surface rigidly fixed to it. The autocollimator is now widely used as a practical tool in precision engineering,[3] and several self-contained instruments are commercially available. It consists of an illuminated wire placed at the focal plane of a lens, as shown in Fig. 2. A plane reflecting surface whose angle of rotation is to be measured is placed on the other side of the lens. The light from the target wire forms a parallel beam between the lens and mirror, and is returned to form an image of the wire in the focal plane of the lens. A small angular movement of the reflecting surface changes the position of the image in the focal plane. The displacement of the image is usually measured with a micrometer eyepiece, calibrated directly in terms of the angular movement of the mirror. On one commercial instrument the micrometer is graduated at intervals of 0·5 sec. of arc, which on a lever of length 2 in. would correspond to a sensitivity of 5×10^{-6} in.

A schematic diagram of a profile measuring instrument using this technique is shown in Fig. 3. The micrometer X is used to traverse the lever arm assembly parallel to the die axis. Vertical displacement of the probe tip, as it follows the bore profile, is compensated by adjustment of the pivot height, using micrometer Y. The mirror plane is thereby restored to its original orientation, as defined by the autocollimator axis. The autocollimator is used as a null indicating device only, and the measurements of the die-profile are read from the micrometer screws.

In view of the greater simplicity of this method in relation to the electrical null detector, an experimental instrument was constructed on these lines.* A photograph of the instrument is shown in Fig. 4. It has been used by the Drawing Section of the Mechanical Working Division of B.I.S.R.A. for three months, in connection with an experimental investigation of die wear, and has been found to give very satisfactory

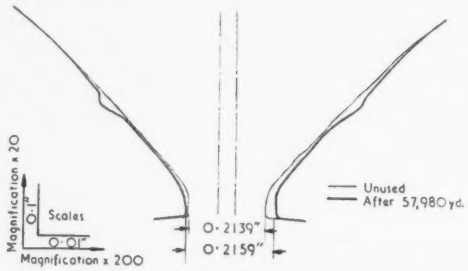

Fig. 5—Plot of wire-drawing-die profile

Fig. 6—Diagram of probe tip (magnified)

results. A typical profile obtained with the instrument during this work is shown in Fig. 5.

The design of the instrument is fully kinematic ; it incorporates high-grade balls-and-needle rollers, which are readily obtained, and avoids precision-ground and lapped beds and surfaces, which are expensive and

* The instrument in this form was demonstrated at the Physical Society's exhibition in 1948.

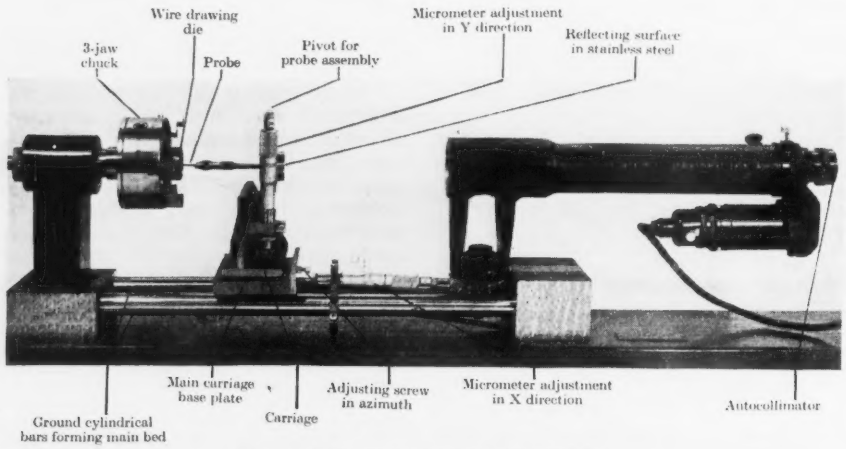

Fig. 4—The experimental instrument

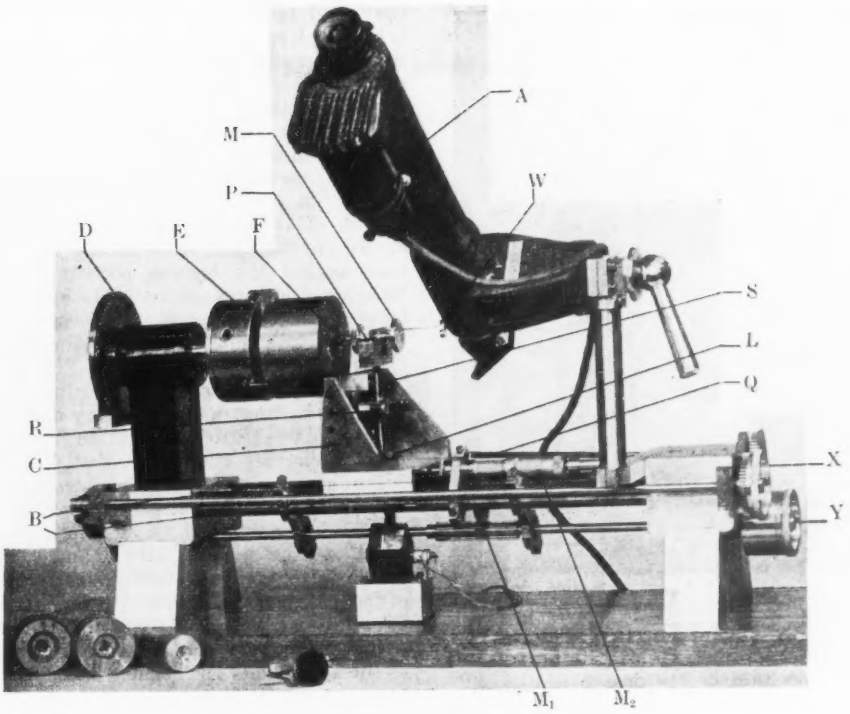

Fig. 7—General view of wire-drawing-die profilometer

difficult to manufacture. A sample probe is shown diagrammatically in Fig. 6 ; the diameter of the tip of a probe of standard size is 0·03 in. For smaller dies, probes with a tip diameter of 0·015 in. are used.

FINAL FORM OF THE INSTRUMENT

From experience gained with the experimental model, one of improved design was built, details of which are given below. Modifications to the design include easier operation and extension of the working range of the instrument.

A photograph of the improved instrument is shown in Fig. 7. The die whose bore profile is to be examined is inserted in the chuck E, which is fitted with a dividing head D, to permit accurate setting of the die to any desired orientation about its own axis, and thus to locate the diametral plane of the profile. The feeler arm F is mounted on a pivot P, the probe tip resting on the inside surface of the die bore, and at the opposite end of the feeler arm a mirror M is fitted, such that any angular rotation of the feeler arm can be detected by the autocollimator A. The feeler-arm pivot P is supported on a vertical spindle S, which can be displaced vertically by the micrometer M_1, which actuates the L-shaped lever L. The pivot, its vertical supporting spindle, adjusting micrometer, and lever assembly are mounted on the carriage C, which slides on the precision-ground silver-steel rods B, so that the whole assembly can be moved parallel to the axis of the die bore. The horizontal translation of the carriage assembly is controlled by the micrometer M_2, acting against the return spring Q. As the micrometer M_2 is adjusted, thus moving the feeler-arm tip axially along the bore of the die, the feeler arm will rotate about its pivot axis. This rotation of the feeler arm is 'nulled' by vertical adjustment of the pivot spindle with the micrometer

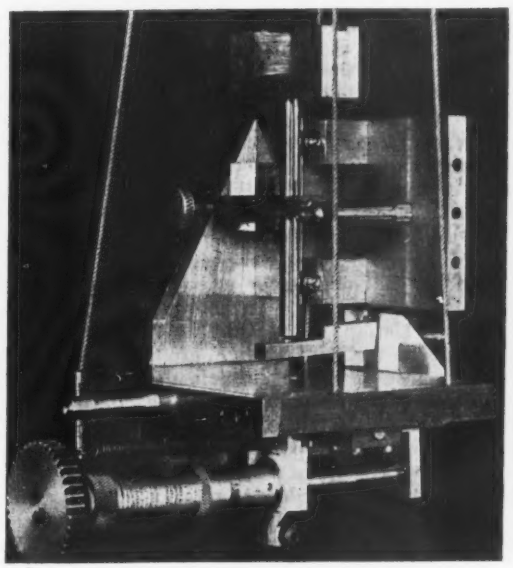

Fig. 8—Feeler-arm carriage

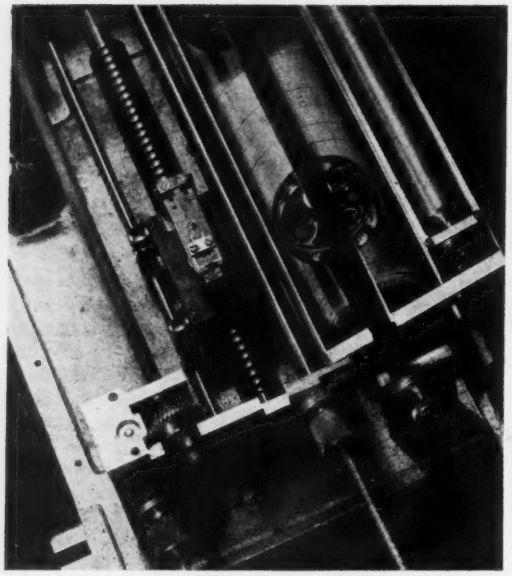

Fig. 9—Plotting table

M_1. This 'null' position is obtained by reference to the initial position of the image in the autocollimator eyepiece. It is an arbitrary zero but, ideally, would correspond to a position such that the plane of the mirror is normal to the axis of the autocollimator. The readings of M_1 and M_2 at the 'null' position give the rectangular co-ordinates of the profile at that position.

For convenience, the micrometers M_1 and M_2 are operated from the adjusting knobs Y and X. The micrometer drives are connected to respective flexible coupling cables, which can be connected to a two-co-ordinate plotting-table recorder, described later. By this means, the curve of the die profile can be plotted as the instrument is adjusted.

A mirror is fitted to the autocollimator, so that the position of the autocollimator can be adjusted to suit the operator's convenience. The autocollimator is mounted on a universal clamp, to facilitate the initial alignment of the optical system.

Figure 8 shows in greater detail the construction of the feeler-arm carriage. To ensure linearity of vertical motion of the feeler-arm pivot spindle with respect to the micrometer screw adjustment, the contacts between the micrometer spindle and one arm of the **L**-shaped lever, and the other arm of the lever and the vertical spindle S, are by balls set into the lever arms bearing against the flats on the pivot spindle and micrometer spindle. The lines joining the axis of rotation of the lever with each of the ball centres are mutually perpendicular.

The feeler arm is mounted on a kinematic pivot, which restricts its freedom to rotation about an axis normal to the length of the feeler arm. The spindle on which this pivot is mounted can be rotated about its own axis by an adjusting screw R. This adjustment permits the tip of the probe to be accurately set in the trough of the die bore. The feeler-arm assembly is balanced, so that the resultant couple is equivalent to a downward force of 1 g. at the tip. The moment of this force is sufficiently small to prevent appreciable bending of the probe, and so to prevent the introduction of any measurable systematic error in the diameter measurement.

When measuring diameters or examining two profiles on opposite ends of a diameter, a rider weight W is used to reverse the resultant couple acting on the lever, thus causing the feeler arm to bear against the highest portion of the inverted trough on the upper surface of the bore. The diameter is given directly as the difference between the micrometer readings, plus a small constant equal to the thickness of the probe tip.

Application of kinematic principles to the design of the instrument was found very appropriate. Care was taken to ensure that wear on the balls and other bearing surfaces would not affect the accuracy.

Performance of the instrument, especially in measuring diameters, depends on a number of precision movements, *e.g.*, lever-arm pivot, feeler-arm pivot, vertical-spindle motion, and carriage motion on the silver-steel rods. Simplicity of construction of the instrument and ease with which manufacturing tolerance can be calculated, while still ensuring the required precision of measurement, are largely due to the adoption of the geometric design.

Plotting Table

The plotting table consists of a pen mechanism capable of linear translation in one direction Y, and a paper chart moving in a direction perpendicular to that of the pen X, the driving mechanisms in the two directions being connected by flexible-cable drives to the two controlling knobs on the instrument.

The pen mechanism is arranged to make a dot on the chart when the operator pushes a button near the control knobs, and a profile can thus be plotted as the adjustments are made, without the operator making records of micrometer readings or removing his eye from the autocollimator eyepiece.

Figure 9 shows the general form of the plotting table.

Acknowledgments

The author would like to thank Mr. S. S. Carlisle, Head of the Instrument Section of the Physics Department of the British Iron and Steel Research Association, for his encouragement and helpful suggestions, and also to acknowledge with thanks the co-operation of Mr. P. H. Freeman who made the instrument.

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Ingot Heat Conservation

TIME STUDIES FROM CASTING TO ROLLING

By A. V. Brancker, Ph.D., J. Stringer, B.A., and L. H. W. Savage, M.Sc., F.R.I.C.

SYNOPSIS

Observations have been made at three works of the progress of hot ingots from the casting bay to the cogging mill, in an attempt to find the causes of delays and their effects on soaking-pit operations. The information obtained from these observations and from a study of the routine records of one works has been analysed to find what improvements are possible, especially in reduction of the time during which ingots can lose heat between teeming and charging to the soaking pit.

It is concluded that worthwhile savings are possible. In the case of one works it is estimated that feasible improvements, including systematic handling of ingots and a new system of keeping records, could result in an increase in the soaking-pit throughput of 18%, made up of cold ingots. Alternatively, if extra ingots were not available, at least 18% of the pit capacity could be taken out of service whilst maintaining the present throughput.

Introduction

THE bulk of the steel made in this country is tapped from open-hearth furnaces and teemed into ingot moulds, the ingots being subsequently stripped and transferred to soaking pits where they are brought to a uniform temperature suitable for rolling.

A recent survey of practice made by the Plant Engineering Division of the British Iron and Steel Research Association as a beginning of an investigation into the basis of mill furnace design, showed that fuel requirements of soaking pits at different works were from 10–20 therms per ton of ingots over long periods, despite the fact that when tapped the steel has a total heat content about 5 therms per ton in excess of that required by the rolling mill.

The present state of knowledge is inadequate to ascertain, other than by trial and error, the optimum times of cooling before stripping, and of heating before rolling. It is impossible to assess what expenditure of fuel is incurred by the provision of a safety factor to ensure that, before stripping, the ingot will have a shell of sufficient strength to be handled without deformation. Without a knowledge of the temperature distribution at the instant of charging the pit, it is also impossible to estimate the heat required, and the time taken for the ingot to attain a sufficiently uniform temperature for rolling.

It is generally accepted, however, that there is a close correlation between track time and minimum heating time, where track time is defined as the interval between the end of teeming and the start of charging to the pits, and heating time as the interval between the end of charging to, and the start of drawing from, the soaking pits. The appropriate correlation varies according to different authorities, *e.g.* :

Heating time = Track time[1]

Heating time = Track time + 1 hr.[2]

but it was noted from the survey referred to above that the average gross heating time varied from about 4½ to 14 hr., compared with minimum track times of about 1½ hr. As in all cases there was a close correlation between fuel consumption and gross heating time, it appeared appropriate to investigate the flow of ingots from the steelmaking plant to the mills, in order to ascertain the factors which increased the track time above the minimum.

This paper describes the observations made at three works, designated *A*, *B* and *C*, where pit casting, bogie or car casting, and ground level casting were the respective practices. A graphical method of plotting the works records has been used for ease in determining the exact position at any time, and in one works by this means the ultimate effect of reducing track and heating times towards attainable minima on the shop performance has been worked out for a complete week, taking as a starting point the actual tapping times of the furnaces.

WORKS *A*—PIT CASTING

Site of Observations

The layout of the No. 2 melting shop and the soaking-pit bay at works *A* is given in Fig. 1. Each of the eight 80-ton open-hearth furnaces has associated with it a ladle car moving on rails above a casting pit. Ingots are teemed in pairs, and are stripped in the casting pits by crane *C*6, which is a special stripper crane, or by one of the two general service cranes, *C*7 or *C*8, using a simple bridle.

The crane used for stripping, loads the ingots on to two cars which are then pushed to the soaking-pit bay by the steam crane *C*9, which is used for miscellaneous duties in the casting bay. Ingots are charged to the pits by cranes *C*4 and *C*5. The track of these cranes overlaps the melting shop as far as the *L* furnace, so that ingots from this furnace can be handled direct from the casting pit to the soaking pits.

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Dr. Brancker, Mr. Stringer, and Mr. Savage are in the Plant Engineering Division of the Association.

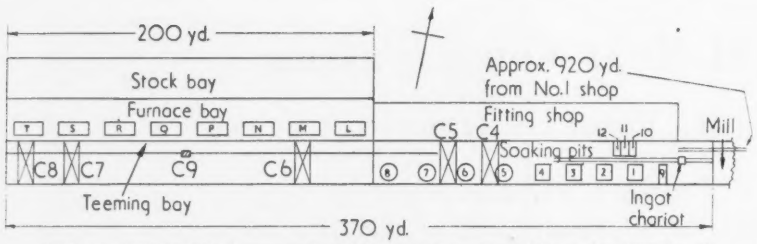

Fig. 1—Schematic layout of melting shop and soaking pits, Works *A*

There are twelve recuperative soaking pits, viz. : Four bottom-fired square pits, of capacity 14 ingots each ; four tangentially fired circular pits, of capacity 13 ingots each ; and four end-fired rectangular pits, of capacity 6 ingots each ; giving a total holding capacity of 132 ingots. The bottoms of coke breeze are cleaned out and remade once or twice weekly and, in general, all twelve pits are in operation. Gas flow is controlled to maintain a preset temperature in the waste-gas ports, and the fuel is automatically shut off when the lids are opened. Ingots are withdrawn by the charging cranes and delivered to the mill racks on a chariot operated by the rolling team. There is a further melting shop with nine 60-ton open-hearth furnaces from which stripped ingots are brought by rail to the soaking-pit bay.

In addition to the ingots cast at weekends and at other times when the soaking pits cannot receive them, a number of ingots are imported. These cold ingots enter the soaking-pit bay by the same road as those from No. 1 melting shop.

Most of the ingots rolled weigh about 6 tons 15 cwt., and are reduced by the cogging mill from an average cross-section of 20½ by 40½ in. to slabs of cross-section 3½ by 36 or 38 in. The average weekly mill output of 10,000–11,000 tons compares with a rated mill capacity of about 13,800 tons per week.

Works Practice

Detailed time observations of the various operations between teeming and rolling were made in the No. 2 melting shop during consecutive days of January, 1948. These operations are considered under the following headings :

Teeming—Twelve to fourteen corrugated moulds are set in pairs in each casting pit, the first pair being under the ladle when the furnace is tapped. Teeming starts with the first pair, and normally proceeds in regular sequence from 1 and 2 to 11 and 12, the last being teemed singly to avoid two stub ends. The slag ladle is in the pit adjacent to the first pair of moulds so that, when a flying nozzle is expected, the order of teeming may be reversed to reduce contamination of teemed ingots by the stream of slag.

The time required to fill one pair of moulds varies from 3–3½ min., and a cast of 11 ingots is made in about 18–20 min., the time taken to move the ladle being only a few seconds.

Solidification Period—The arbitrary rule followed is that the steel should be left for a period of 1¼ to 1½ hr. to acquire a sufficiently robust shell before stripping. Periods of less than 1¼ hr. are not uncommon. There is a considerable variation between the solidification periods of individual ingots, as will be shown later (*see* also Fig. 2).

Stripping—Ingots are normally stripped by the special stripper crane fitted with a ram. The crane-driver has a restricted field of vision, and it is more convenient for him to strip first the ingots furthest from the tap hole. In general, this means that stripping follows the reverse order of teeming, and the solidification times of individual ingots may vary by as much as 25 min. The time taken to remove the mould, to deposit it on the bank and to return to strip the next ingot is about 1½ min., so that the stripping of a complete cast takes 15–20 min. The stripper ram is too short to loosen incompletely filled, sticking, moulds. Such moulds must be placed on the bank and a block inserted between the stripper ram and the ingot, which may delay the stripping operation by as much as 10 min.

When a general service crane is used instead of the stripper crane the operation is similar, except that sticking moulds must be freed by impact with stationary moulds or be set aside to cool.

Transfer of Ingots from Casting Pit to Soaking Pits—The two ingot cars are usually transferred to the soaking-pit bay by the steam crane. As there is only a single track, and this crane may be engaged on other duties, it is not always practicable for the cars to be brought to the pitside before or during the stripping operation to enable the ingots to be loaded at the first opportunity. The time taken in loading is about ½ min. for each ingot and 5–6 min. for a complete cast. To avoid interference between the cranes, loading is not normally begun until all the ingots have been stripped, and does not always follow any regular sequence.

The journey to the soaking pits takes up to 3 min. after the steam crane becomes available, depending on the distance to be travelled. At the first opportunity the ingots are transferred into a soaking pit by one of the pit cranes. The operation requires about ½–1 min. per ingot, and if one of the cranes is fully available a complete cast will be transferred from the cars in 5–10 min. Frequently, a single crane is both feeding the mill from one pit and charging a second, and the time to transfer a cast from the cars may be extended to 15 or 20 min. If a pit is recharged while being drawn the transfer time is extended to about 40 min.

Soaking and Heating Period—Ingots are normally considered in four broad categories, namely, hot, black hot, warm, and cold. On these are based minimum heating times approximating to the formula :

$$\text{Heating time} = \text{Track time plus 1 to } 1\tfrac{1}{2} \text{ hr.}$$

However, due to a large overall holding capacity of the pits, two or more casts may frequently be ready for drawing at one time and, in consequence, one or more of them must be left soaking for longer than the minimum time.

Drawing and Rolling—The operation of drawing an ingot and placing it on the chariot for transfer to the mill racks requires the use of a crane for about ½ min. The rate of drawing, about one every 3–3½ min., is governed by the mill, which can reduce an ingot in about 2½ min. A complete cast is drawn in about 35 min. if no breaks occur. The order in which individual ingots are drawn is not necessarily the same as that in which they were charged.

The casting bay operations in the No. 1 melting shop are similar to those described, except that because of the smaller furnaces the average cast is only 9 ingots, and no stripper crane is used. The moulds are lifted by an E.O.T. crane and, if the ingots do not slip out, are transferred to a ground stripper. On completion of stripping, the ingots are loaded on to bogies and transferred by a locomotive to the soaking pits. Since a locomotive is not always available the track time may vary considerably, from 1¾ hr. upwards, as is shown by the works records, the mean over a full week being 3·62 hr.

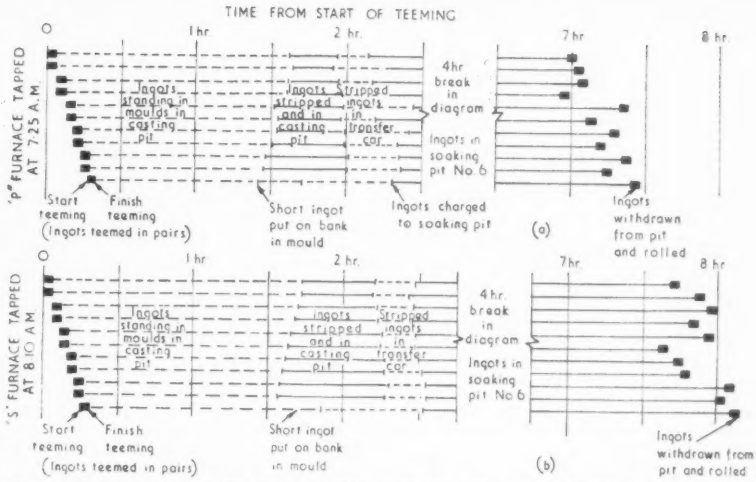
TIME FROM START OF TEEMING
'P' FURNACE TAPPED AT 7.25 A.M.
Ingots standing in moulds in casting pit
Ingots stripped and in casting pit
Ingots stripped in transfer car
4hr break in diagram
Ingots in soaking pit No 6
Start teeming
Finish teeming
(Ingots teemed in pairs)
Short ingot put on bank in mould
Ingots charged to soaking pit
Ingots withdrawn from pit and rolled
(a)
'S' FURNACE TAPPED AT 8.10 A.M.
(b)

Fig. 2—Detailed time study on ingots from two casts at Works *A*. Pit casting (*a*) cast P683; (*b*) cast S369 (eleven 6¾-ton ingots)

Observed Data

Details of the movements of the individual ingots

Table I

TIME STUDY OBSERVATIONS ON CAST FROM *P* FURNACE, TAPPED AT 7.25 A.M. (ZERO TIME) AND *S* FURNACE, TAPPED AT 8.10 A.M. (ZERO TIME)—WORKS *A*

Ingot	Finished Teeming	Stripped	Loaded to Car	Charged to Soaker	Drawn from Soaker	Teeming Time	Time in Mould	Stripping to Charging	Track Time	Time in Pit	Heating Time*	Heating Time minus Track Time
	Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Min.	Hr. Min.	Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min
1	9	1 43½	2 1	2 12	7 2½	3½	1 34½	28½	2 3	4 50½	4 31	2 28
2	9	1 42½	2 0	2 20½	7 5½	3½	1 33½	38	2 11½	4 45	4 34	2 22½
3	13	1 41	2 2	2 4½	7 8½	3¼	1 28	23½	1 51½	5 4	4 37	2 45½
4	13	1 40	2 1½	2 18½	6 59½	3¼	1 27	38½	2 5½	4 41	4 28	2 22½
5	16	1 39	2 3½	2 30½	7 24	3	1 23	51½	2 14½	4 53½	4 52½	2 38
6	16	1 37	2 3	2 15½	7 11½	3	1 21	38½	1 59½	4 56	4 40	2 40½
7	19	1 35	2 5	2 31½	7 21½	3	1 16	56½	2 12½	4 50	4 50	2 37½
8	19	1 34	2 4½	2 24½	7 14½	3	1 15	50½	2 5½	4 50	4 43	2 37½
9	22	1 32	2 6½	2 25½	7 26½	3¼	1 10	53½	2 3½	5 1	4 55	2 51½
10	22	1 31	2 6	2 27½	7 18	3¼	1 9	56½	2 5½	4 50½	4 47	2 41½
12	25	1 47	1 59	2 22	7 30½	3	1 22	35	1 57	5 8½	4 59	3 2
(Short Ingot) Track time of whole cast, 1 hr. 39½ min. Heating time of whole cast, 4 hr. 28 min.						Delay from last ingot stripped to first ingot loaded on car, 12 min.						
1	9	1 48	2 17	2 33	7 26	3½	1 39	45	2 24	4 53	4 48	2 24
2	9	1 47	2 16	2 31	7 57	3½	1 38	44	2 22	5 26	5 19	2 57
3	12	1 46	2 19	2 30	7 54	3	1 34	44	2 18	5 24	5 16	2 58
4	12	1 45	2 18	2 32	8 2	2¾	1 33	47	2 20	5 30	5 24	3 4
5	15	1 43	2 20	2 35	7 51	3	1 28	52	2 20	5 16	5 13	2 53
6	15	1 42	2 19½	2 29	7 43	3	1 27	47	2 14	5 14	5 5	2 51
7	18	1 41	2 22	2 34	7 48	2¾	1 23	53	2 16	5 14	5 10	2 54
8	18	1 40	2 21	2 32	8 0	3	1 22	52	2 14	5 28	5 22	3 8
9	21	1 39	2 23	2 37	8 8	2¾	1 18	58	2 16	5 31	5 30	3 14
10	21	1 38	2 22½	2 38	8 5	2¾	1 17	60	2 17	5 27	5 27	3 10
12	23½	1 55	2 12	2 36	8 11	2½	1 26½	41	2 12½	5 35	5 33	3 20½
(Short Ingot) Track time of whole cast, 2 hr. 5½ min. Heating time of whole cast, 4 hr. 48 min.						Delay from last ingot stripped to first ingot loaded on car, 17 min.						

* Taken from time of charging of last ingot.

of two casts from No. 2 melting shop are given in Table I, and are illustrated in Fig. 2.

The track time of an individual ingot has been taken as the interval between the completion of teeming and the start of charging to the pit. The heating time, however, has been assumed not to begin until all the ingots of the cast have been charged, as during the charging operation the fuel is shut off while the lids are open, and there is no factual evidence to substantiate whether there is a net gain or loss of heat from ingots in the pit. When drawing for the mill, the fuel is off for only about ½ min. at 3–3½ min. intervals, and little error should arise in assuming that the heating period for an individual ingot continues until the moment it is drawn.

It may be seen that (i) the track times and the heating times of the individual ingots differ considerably from those of the cast, (ii) the practice of stripping in the reverse order to that of teeming results in the majority of the ingots having a longer solidification time than necessary, (iii) any delays have an adverse effect on the whole cast.

Possible Systematic Operations

It is pertinent to consider the effect of smooth operation, free from delays, on both the normal practice as discussed above, and on a modified practice in which attention is particularly given to ensure that individual ingots are handled at the optimum time. In order to do this, it has been necessary to make the following assumptions, which observations suggest are readily attainable, but are by no means minima :

Teeming—Teeming time per pair of ingots is 3½ min. A cast of either 11 or 12 ingots will be completely teemed 26½ min. after commencement of tapping the furnace.

Solidification Period—Solidification time is 1 hr. 9 min.

Table II
SYSTEMATIC INGOT HANDLING

Practice	Ingot No.	Teemed	Ready for Stripping	Stripped	Loading on Car*	Charged	Track Time	Ready to Draw	Drawn (1)	Drawn (2)
		Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.
A(1) and (2)	1	9	1 18	1 51		2 0	1 51	5 31	5 31	5 58½
	2	9	1 18	1 49½		2 1	1 52	5 32	5 34½	5 55
	3	12½	1 21½	1 48		2 2	1 49½	5 29½	5 38	5 51½
	4	12½	1 21½	1 46½		2 3	1 50½	5 30½	5 41½	5 48
	5	16	1 25	1 45	1 52	2 4	1 48	5 28	5 45	5 44½
	6	16	1 25	1 43½	to	2 5	1 49	5 29	5 48½	5 41
	7	19½	1 28½	1 42	1 57	2 6	1 46½	5 26½	5 52	5 37½
	8	19½	1 28½	1 40½		2 7	1 47½	5 27½	5 55½	5 34
	9	23	1 32	1 39		2 8	1 45	5 25	5 59	5 30½
	10	23	1 32	1 37½		2 9	1 46	5 26	6 2½	5 27
	11	26½	1 35½	1 36		2 10	1 43½	5 23½	6 6	5 23½
A(3)	1					2 10	2 1	5 41	5 48½	
	2					2 9	2 0	5 40	5 45	
	3					2 8	1 55½	5 35½	5 41½	
	4					2 7	1 54½	5 34½	5 38	
	5					2 6	1 50	5 30	5 34½	
	6		as above			2 5	1 49	5 29	5 31	
	7					2 4	1 44½	5 24½	5 27½	
	8					2 3	1 43½	5 23½	5 24	
	9					2 2	1 39	5 19	5 20½	
	10					2 1	1 38	5 18	5 17	
	11					2 0	1 33½	5 13½	5 13½	
B(1) and (2)	1	9	1 18	1 21		1 46	1 37	5 3	5 3	5 30½
	2	9	1 18	1 22½		1 47	1 38	5 4	5 6½	5 27
	3	12½	1 21½	1 24		1 48	1 35½	5 1½	5 10	5 23½
	4	12½	1 21½	1 25½		1 49	1 36½	5 2½	5 13½	5 20
	5	16	1 25	1 27	1·37	1 50	1 34	5 0	5 17	5 16½
	6	16	1 25	1 28½	to	1 51	1 35	5 1	5 20½	5 13
	7	19½	1 28½	1 30	1·42	1 52	1 32½	4 58½	5 24	5 9½
	8	19½	1 28½	1 31½		1 53	1 33½	4 59½	5 27½	5 6
	9	23	1 32	1 33		1 54	1 31	4 57	5 31	5 2½
	10	23	1 32	1 34½		1 55	1 32	4 58	5 34½	4 59
	11	26½	1 35½	1 36		1 56	1 29½	4 55½	5 38	4 55½
B(3)	1					1 56	1 47	5 13	5 20½	
	2					1 55	1 46	5 12	5 17	
	3					1 54	1 41½	5 7½	5 13½	
	4					1 53	1 40½	5 6½	5 10	
	5					1 52	1 36	5 2	5 6½	
	6		as above			1 51	1 35	5 1	5 3	
	7					1 50	1 30½	4 56½	4 59½	
	8					1 49	1 29½	4 55½	4 56	
	9					1 48	1 25	4 51	4 52½	
	10					1 47	1 24	4 50	4 49	
	11					1 46	1 19½	4 45½	4 45½	

* Order of loading is immaterial as long as the ingots can be identified.

Stripping Time—Stripping time is 1½ min. per ingot. No additional time has been allowed for stripping a short sticking ingot. Delays due to this could be avoided in at least three ways :

(i) By extending the useful stroke of the stripping ram

(ii) By using a second crane for loading ingots while the short ingot is being stripped. It may be noted that in normal circumstances the use of a second crane to load the car, while the stripper crane is depositing the mould, would enable the ingot car to move off 1 min. after the last ingot is stripped, giving a further potential saving of 5 min. in the track time

(iii) By not attempting to strip the short ingot until all the other ingots have been loaded on to the ingot car. The short ingot could then be stripped and carried down the shop by the crane.

Car Loading—Beginning immediately the cast is stripped, car loading time is ½ min. per ingot.

Journey to Soaking Pits—The journey to the soaking pits occupies 3 min.

Charging Soakers—Charging soakers occupies 1 min.

Heating Time—Heating time is a minimum of individual track time plus 1½ hr., beginning immediately the last ingot is charged.

Drawing Time—Drawing time is 1 ingot every 3½ min.

Details for the individual ingots have been worked out in Table II for six schemes, namely :

(A) Ingots stripped in reverse order of teeming :
  (1) Charged and drawn in order of teeming
  (2) Charged in order of teeming, drawn in reverse order
  (3) Charged and drawn in reverse order of teeming.
(B) Ingots stripped in order of teeming, for the same practices (1), (2), and (3).

The normal order of operations in this works is represented by scheme A(1).

The overall effects on the cast operations are summarized in Table III, in which the corresponding data from Table I have been included for comparison. These tables suggest that the order in which the ingots are taken is of potential importance. The arbitrary heating rule, based on total track time, will be influenced by the proportion of its track time that an ingot is in a stripped condition. Nevertheless, reference to Fig. 3, on which the cooling of the steel is shown graphically, indicates that, with the possible exception of practice B(3) this rule should be just as applicable as for the present normal practice, which is approximately represented by A(1).

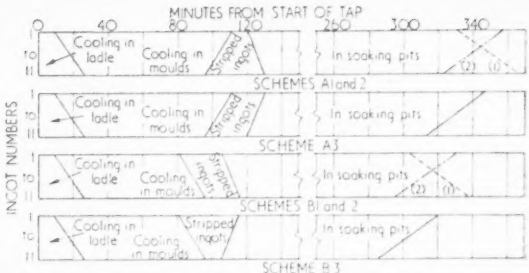

Fig. 3—Illustration of systematic methods of ingot handling, Works *A*

Each of the modified practices requires the use of the soaking pits for a shorter time than the delay-free current practice, and therefore tends to make it easier to transfer subsequent casts in the minimum track time.

However, as the open-hearth furnaces cannot yet be tapped to a regular schedule, it may not be possible to take full advantage of this potential saving of time, unless the operations can be so arranged that sufficient consecutive time intervals are available on one pit to heat an additional cast. In a later section of the report a full week's operation, based on actual tapping times, has been worked out to examine the long-term effects of reducing track times.

WORKS *B* —BOGIE CASTING

Site of Observations

The disposition of plant at the site of observations is schematically illustrated for Works *B* in Fig. 4, and comprises 7 open-hearth furnaces, one mixer, 5 end-fired reversing pits, and 6 one-way-fired recuperative pits.

At this works the steel is tapped into a ladle of 90 tons capacity, which is handled by one of the ladle cranes *C*1, *C*2, or *C*3. A train of some eight bogies, each carrying four ingot moulds set in single file, is marshalled alongside one of the teeming platforms *T*1, *T*2, *T*3, by a locomotive. The steel throughput is about 5000 tons per week, and most of this is cast as 3-ton ingots, but slabbing ingots of 5 tons weight are also made from time to time.

Table III

EFFECT OF SYSTEMATIC INGOT HANDLING AT WORKS *A*

Operation	Table I		Systematic Operations (Table II)					
	P Furnace	*S* Furnace	A(1)	A(2)	A(3)	B(1)	B(2)	B(3)
	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.
Start tap to finish of teeming ...	25	23½	26½	26½	26½	26½	26½	26½
Finish teeming to start charging (track time)...	1 39½	2 5½	1 33½	1 33½	1 33½	1 19½	1 19½	1 19½
Charging soakers	27	9	10	10	10	10	10	10
Finish charging to start drawing (heating time)	4 28	4 48	3 21	3 13½	3 3½	3 7	2 59½	2 49½
Drawing	31	45	35	35	35	35	35	35
Total, start tap to finish drawing	7 30½	8 11	6 6	5 58½	5 48½	5 38	5 30½	5 20½
Total time pit occupied ...	5 26	5 41	4 6	3 58½	3 48½	3 52	3 44½	3 34½

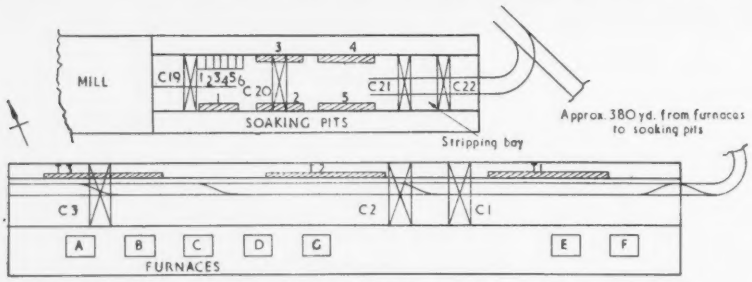

Fig. 4—Schematic layout of melting shop and soaking pit, Works *B*

On completion of teeming, the bogies are hauled by locomotive from the casting to the stripping bay in two trains, each holding some 12 to 16 ingots. In the stripping operation, crane *C*21 or *C*22 is used, and the existing practice is to strip first those ingots nearest to the soaking pits. Consequently, because of the shunt necessary in the transfer of the moulds from the casting to the stripping bay, the stripping order is the reverse of the teeming order; reference to this procedure will be made again later. Each ingot, after stripping, is charged with minimum delay to the soaking pits by crane *C*19 or *C*20.

Each of the 5 end-fired soaking pits has a capacity of 20 ingots, whilst the 6 one-way-fired recuperative pits hold 6 ingots each. Since one pit of each type is, in general, under repair the normal holding capacity is 110 ingots of 3 tons each. When the ingots have attained rolling temperature, they are drawn from the pits by crane *C*19 or *C*20 and are placed on a chariot for transfer to the rolling mill.

Observed Data

In the track time study made at Works *B*, a complete cast of 28 ingots was followed from casting to rolling, and observations were made on the handling and transit times of individual ingots. The data thus obtained are summarized in Table IV, and are graphically illustrated in Fig. 5.

The ingots, of 3 tons weight, were teemed at a rate of approximately 1 ingot per min., and in this operation teeming proceeded from ingot 1 to 28, where ingot 28 was the furthest from the locomotive end of the train of bogies. The locomotive arrived some 6½ min. after the last ingot was teemed, and hauled a batch of 16 ingots to the stripping bay. It then returned, and a second batch of 12 ingots was hauled to the stripping bay. The times taken for these two journeys were 3 and 4 min. respectively, and no undue delays in shunting were observed.

Figure 5 illustrates the fact that, because of the shunt system, ingot No. 16 was the nearest to the soaking pits, and was therefore the first to be handled. In other words, it will be observed that the stripping

Table IV—TIME STUDY OBSERVATIONS ON ONE CAST AT WORKS *B*
Furnace Tapped at 10.58 a.m. (zero time)

Ingot	Finish Teeming	In Transit	Stripped	Charged	Drawn	Pit*	Time in Mould	Track Time	Heating Time†	Heating Time minus Track Time
	Min.	Min.	Hr. Min.	Hr. Min.	Hr. Min.		Hr. Min.	Hr. Min.	Hr. Min	Hr. Min.
1	10½		1 3	1 5	5 11	*b*	52½	54½	3 52	2 57½
2	11½		1 5	1 7	5 5	*b*	53½	55½	3 46	2 50½
3	12½		1 15	1 16	4 29	*b*	1 2½	1 3½	3 10	2 6½
4	14		1 2	1 3	5 8	*b*	48	49	3 51	3 2
5	15		1 20	1 21	4 35	*b*	1 5	1 6	3 16	2 10
6	16		1 19	1 20	4 44	*b*	1 3	1 4	3 25	2 21
7	17		1 11	1 11	4 54	*b*	54	54	3 35	2 41
8	18	46 to 49	1 9	1 10	4 57	*b*	51	52	3 38	2 46
9	19		1 8	1 9	5 0	*b*	49	50	3 41	2 51
10	20		1 1	1 6	5 18	*b*	41	46	3 59	3 13
11	21		59	59	3 34	*a*	38	38	2 35	1 57
12	22½		57	57	3 31	*a*	34½	34½	2 32	1 57½
13	23½		55	55	3 28	*a*	31½	31½	2 29	1 57½
14	24½		54	1 1	5 15	*b*	29½	36½	3 36	2 59½
15	25½		52	52	3 25	*a*	26½	26½	2 36	2 9½
16	26½		49	50	3 22	*a*	22½	23½	2 23	1 59½
17	27½		1 30	1 30	3 37	*c*	1 2½	1 2½	1 59	56½
18	28½		1 16	1 17	4 5	*b*	47½	48½	2 27	1 38½
19	30		1 13	1 34	3 43	*c*	43	1 4	2 5	1 1
20	31		1 29	1 31	3 49	*c*	58	1 0	2 11	1 11
21	32		1 25	1 32	3 40	*c*	53	1 0	2 2	1 2
22	33	53 to 57	1 23	1 37	3 46	*c*	50	1 4	2 8	1 4
23	34		1 21	1 38	3 52	*c*	47	1 4	2 14	1 10
24	35		1 20	1 23	4 1	*b*	45	48	2 23	1 35
25	36		1 7	1 12	4 21	*b*	31	36	3 2	2 26
26	37		1 17	1 19	4 9	*b*	40	42	2 50	1 8
27	38½		1 12	1 18	4 8	*b*	33½	39½	2 49	1 9½
28	39½		1 6	1 8	5 2	*b*	26½	28½	3 43	3 14½

* Pits *a* and *c*—One-way-fired recuperative, capacity 6 ingots, *b*—Regenerative, capacity 20 ingots.
† Taken from the moment of charging the last ingot to each pit.

order did not follow the teeming order. From Table IV the time intervals between teeming and stripping vary from 22½ min. to 1 hr. 5 min., and the cause of the scatter in these times can be traced to the stripping order employed.

Possible Systematic Operations

Table V gives data for two hypothetical cases where ingots are handled systematically. In the first of these two examples, the ingots, after teeming, are hauled to the stripping bay in batches of 16 and 12, at the respective times of 28 and 34 min. after the start of the teeming, the locomotive making two successive journeys.

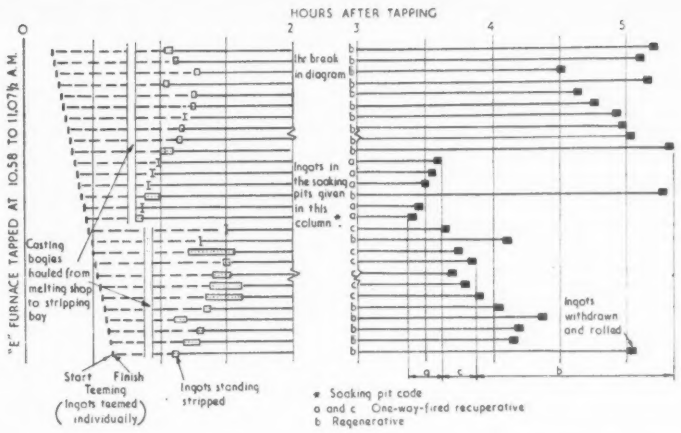
HOURS AFTER TAPPING
"E" FURNACE TAPPED AT 10.58 TO 11.07 h A.M.
the break in diagram
Ingots in the soaking pits given in this column *
Casting bogies hauled from melting shop to stripping bay
Start Teeming
Finish Teeming
(Ingots teemed individually)
Ingots standing stripped
Ingots withdrawn and rolled
* Soaking pit code
a and c One-way-fired recuperative
b Regenerative

Fig. 5—Detailed time study on one cast of ingots at Works *B*. Car casting, cast 156E (twenty-eight 3-ton ingots)

Stripping then occurs in the same order as teeming, namely, from ingot 1 to ingot 16, and from ingot 17 to ingot 28, at the rate of 1 ingot per min. On this basis, stripping commences at 32 min. and ends at 59 min., and as the ingots are stripped they are immediately charged to the soaking pits at a rate of 1 every minute. This charging process therefore starts at 33 min. and finishes at 60 min.; or the entire process of teeming, stripping, and charging 28 ingots takes exactly 1 hr.

It is then assumed that, for the first ingot charged at 33 min., the soaking time equals track time plus 1½ hr., which is 15 min. longer than the minimum time observed, or ingot 1 is drawn at 2 hr. 35 min., and ingots 2 to 28 are drawn thereafter without any breaks, at successive intervals of 3 min., whereby

Table V

SYSTEMATIC OPERATIONS AT WORKS *B*—TIMES TAKEN FROM START OF TEEMING

Finish Teeming	Example 1						Example 2					
	In Transit	Stripped*	Drawn	Time in Mould	Track Time	Soaking Time	In Transit	Stripped*	Drawn	Time in Mould	Track Time	Soaking Time
Min.	Min.	Min.	Hr. Min.	Min.	Min.	Hr. Min.	Min.	Min.	Hr. Min.	Min.	Min.	Hr. Min.
1		32	2 35			2 2		20	2 11			1 50
2		33	2 38			2 4		21	2 14			1 52
3		34	2 41			2 6		22	2 17			1 54
4		35	2 44			2 8		23	2 20			1 56
5		36	2 47			2 10		24	2 23			1 58
6		37	2 50			2 12		25	2 26			2 0
7		38	2 53			2 14		26	2 29			2 2
8	28	39	2 56			2 16	16	27	2 32			2 4
9	to	40	2 59			2 18	to	28	2 35			2 6
10	31	41	3 2			2 20	19	29	2 38			2 8
11		42	3 5			2 22		30	2 41			2 10
12		43	3 8			2 24		31	2 44			2 12
13		44	3 11			2 26		32	2 47			2 14
14		45	3 14			2 28		33	2 50			2 16
15		46	3 17	31	32	2 30		34	2 53	19	20	2 18
16		47	3 20			2 32		35	2 56			2 20
17		48	3 23			2 34		36	2 59			2 22
18		49	3 26			2 36		37	3 2			2 24
19		50	3 29			2 38		38	3 5			2 26
20		51	3 32			2 40		39	3 8			2 28
21		52	3 35			2 42	28	40	3 11			2 30
22	34	53	3 38			2 44	to	41	3 14			2 32
23	to	54	3 41			2 46	31	42	3 17			2 34
24	37	55	3 44			2 48		43	3 20			2 36
25		56	3 47			2 50		44	3 23			2 38
26		57	3 50			2 52		45	3 26			2 40
27		58	3 53			2 54		46	3 29			2 42
28		59	3 56			2 56		47	3 32			2 44

* Charged to soaking pit 1 min. after stripping.

drawing finishes at 3 hr. 56 min. In other words the entire sequence of operations for the 28 ingots from commencement of teeming to the end of drawing takes 3 hr. 56 min.

In the absence of temperature measurements, it is not possible to state with certainty what is a safe minimum solidification time to allow before stripping, but it is of interest to consider the effect of shortening this time. In the second example, therefore, the solidification time has been arbitrarily reduced from 31 to 19 min., the shortest time observed was 22½ min., and the procedure in this second example follows that assumed for the first, but with the difference that the first batch of 16 ingots is hauled to the stripping bay at 16 min., instead of at 28 min., whilst the second batch of 12 ingots is hauled at 28 min. instead of 34 min. In other words, the transfer of ingots to the stripping bay is not delayed until all 28 ingots are teemed, but proceeds after No. 16 is teemed.

As before, the teeming, transit, stripping, charging, and drawing rates are respectively 1, 3, 1, 1, and 3 min. per ingot, and the soaking time for ingot 1 conforms to the same rule as that used in example 1.

The main features derived from a study of these two examples are summarized and compared with the observed data in Table VI.

Although the two examples considered are hypothetical, the results obtained demonstrate fairly clearly the advantages to be gained by adhering to systematic handling, as distinct from irregular handling, sequences. Moreover, a better control of soaking times is obtained by using a regular handling sequence.

In Table V the soaking times increase regularly from ingot 1 to ingot 28, and in both examples studied the difference between the first and last ingot is some 54 min. The factor responsible for this steady increase in soaking time is the rate of drawing or rolling, which has been put at 1 ingot per 3 min.

In practice, the conditions assumed for the two hypothetical cases may not always be realized, and a sequence of operations may be interrupted for various lengths of time, from many adventitious causes. Therefore, it is not suggested that the reduction in overall operating times established for the two hypothetical cases can, in practice, be obtained simply by following a regular handling sequence, but such sequence, if approached whenever possible, would result in a saving in time and a conservation of heat.

The final aspect to be considered in this study of track times at Works *B*, is that, between the casting and stripping bays, moulds and ingots are subjected to variable atmospheric conditions which may induce rapid cooling. As has been stated, transit times observed are of the order of 3 min., but it may be suspected that the times will, on occasions, be even longer, owing to delays. In this case, heat losses may be unduly high, although they cannot be estimated until temperature measurements have been made.

Table VI

DATA FROM HYPOTHETICAL EXAMPLES COMPARED WITH OBSERVED DATA

Example	Completion of Teeming Min.	Sec.	Stripping Hr.	Min.	Charging Hr.	Min.	Drawing Hr.	Min.	Track Time
(observed)	28	48	1	20	1	27	5	7	27 min. to 1 hr.7min.
1	28	0	0	59	1	0	3	56	32 min.
2	28	0	0	47	0	48	3	32	20 min.

WORKS *C*—GROUND LEVEL CASTING

Site of Observations

The layout of the No. 2 melting shop and the casting and soaking-pit bays at Works *C* is shown in Fig. 6.

This melting-shop contains five 90-ton fixed furnaces, one 300-ton tilting furnace, and a 350-ton mixer. Hot metal is poured from the casting side either to the mixer or into a transfer ladle running below the stage to supply the furnaces directly. Normally, the hot-metal ladles enter the shop from the west and have only to be brought past the tilting furnace. However, at the time the observations were made, it was temporarily necessary to bring the hot-metal ladles in at the east end, and to convey them past the fixed furnaces.

The transfer of hot metal from the mixer to the furnaces occurs entirely on the charging side, and does not interfere with casting-bay operations.

The casting bay contains four heavy ladle cranes and two light general-service cranes on a higher gantry. Ingots are teemed in pairs from 90-ton ladles from teeming platforms on the north side of the casting bay, the moulds standing in a single line on base plates at ground level. Steel is normally cast into slabbing ingots of about 6½ tons weight, with an average cross-section of approximately 40½ in. by 21 in., about 12 ingots being obtained from each cast of the fixed furnaces, and from each of the three full ladles normally tapped from the tilting furnace. Each half of the bay is divided into four teeming sites, each containing 12 or 14 standard moulds and 4 or 2 smaller moulds, the latter being used towards the end of teeming, if necessary, to avoid short ingots. A small proportion of ingots are cast in different sized moulds, some with hot tops, mainly for export to other works.

The soaking-pit bay, continuing on to the mills, is at right angles to the casting bay, and contains four

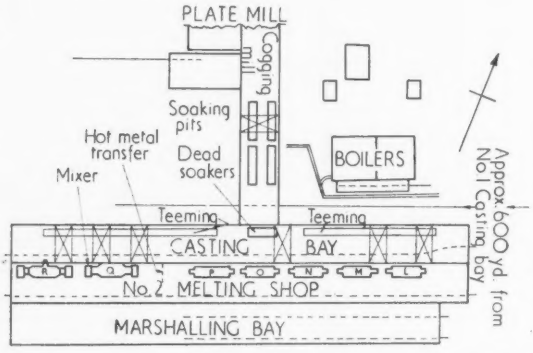

Fig. 6—Schematic layout of melting shop and soaking pits, Works *C*

end-fired regenerative soaking pits, each of 12 ingots capacity. Three of the pits are divided into 12 cells, each with a plug cover, while the fourth pit is a single chamber with rolling covers. The crane gantry extends into the casting bay, so that both the single soaking-pit crane and the casting-bay cranes can operate at the junction where there are four dead soakers, each with a capacity of 5 ingots.

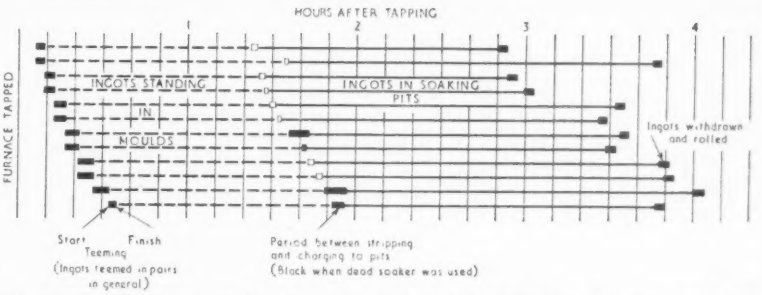

Fig. 7—Detailed time study on one cast of ingots at Works *C*. Ground-level casting, cast from *M* furnace (twelve 6½-ton ingots)

In the No. 1 melting shop, which contains three 80-ton furnaces, the ingots are teemed in pits. After stripping they are transferred to bogies, and conveyed by locomotive to the soaking pits by the track running to the north of the casting bay.

Ingots are rolled in the slabbing mill and, after shearing into lengths, are normally transferred immediately to reheating furnaces before supplying the plate mill. As it is not desirable to produce more slabs than can be accommodated in the reheating furnaces, at times when there is an ample supply of ingots ready for drawing from the pits, the main products are heavy plates, only 2 slabs being cut from one ingot. Conversely, if there is a shortage of soaked ingots, an ingot will be reduced to 5 or 6 slabs, which will be rolled to light plates and will occupy the plate mill for a longer period. Normally the only cold ingots charged to the soaking pits are those arising from weekend casts.

Observed Data

The usual ingot-handling practice is as follows: After a solidification period of about 1–1¼ hr., ingots are stripped by one of the ladle cranes, using a simple bridle to lift the mould. Occasional 'stickers' are dislodged by swinging against the adjacent mould. While the empty mould is being deposited in the middle of the bay to cool, the stripped ingot is picked up by one of the light cranes and carried to the junction of the casting and soaking-pit bays. If a pit is available the ingot is merely lowered to the ground and picked up by the soaking-pit crane, but if all the pits are occupied, it is placed immediately in one of the dead soakers. During this operation the next ingot has been stripped, and the process is repeated.

The data obtained on one complete cast of ingots from one of the fixed furnaces between teeming and rolling, at a period when there was no interference by other casting-bay operations, are given in Table VII, and shown graphically in Fig. 7.

It may be observed that with the exception of the second ingot, initially a 'sticker,' the stripping order followed the teeming order, so that the solidification periods were substantially the same in every case. In the majority of cases the synchronization of the cranes was such that the ingot was charged to the soaking pits within 2 min. of stripping. On two occasions, however, when the soaking-pit crane was feeding the mill, the interval was extended to 7 and 8 min. respectively, but the ingots were placed in the dead soaker to reduce the loss of heat.

The order of drawing from the pits approximately followed the charging order, the first ingot being drawn after a heating time only slightly longer than the track time. However, owing to the slower rate of rolling compared with the rate of charging, coupled with a slight delay at a change of shifts, subsequent ingots were necessarily soaked for increasing periods,

Table VII

TIME STUDY OBSERVATIONS ON ONE CAST—WORKS *C*

Furnace Tapped at 10.40 a.m. (zero time)

Mould No.	Finish Teeming	Stripped*	Dropped at Dead Soaker*	Charged to Live Soaker*	Drawn*	Time in Mould	Track Time	Time in Pit	Heating Time minus Track Time
	Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.	Hr. Min.
1	9½	1 22	1 23	1 24	2 50	1 12½	1 14½	1 26	11½
2	9½	1 34	1 35	1 36	3 45	1 24½	1 26½	2 9	42½
3	12½	1 25	1 25	1 27	2 54	1 12½	1 14½	1 27	12½
4	12½	1 27	1 27	1 28	3 0	1 14½	1 15½	1 32	16½
5	17	1 29	1 30	1 31	3 32	1 12	1 14	2 1	47
6	17	1 32	1 32	1 33	3 39	1 15	1 16	2 6	50
7	21½	1 36	1 37	1 43	3 57	1 14½	1 21½	2 14	52½
8	21½	1 41	1 41	1 42	3 51	1 19½	1 20½	2 9	48½
9	27	1 43	1 44	1 45	4 13	1 16	1 18	2 28	1 10
10	27	1 46	1 47	1 48	4 18	1 19	1 21	2 30	1 9
11	32½	1 49	1 51	1 57	4 38	1 16½	1 24½	2 41	1 16½
12	35	1 52	1 52	1 56	4 23	1 17	1 21	2 27	1 6

* To the nearest minute.

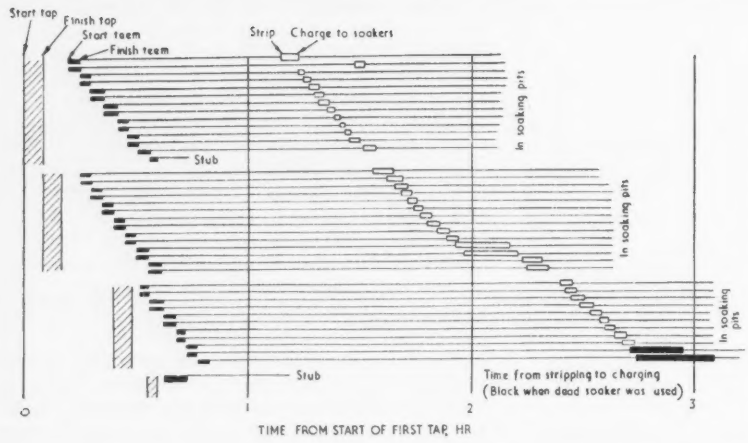

Fig. 8—Tapping, teeming and stripping cast from tilting furnace, Works *C*

amounting, for the last ingot, to track time plus 1 hr. 17 min.

At least three full ladles can be tapped from the tilting furnace, causing a certain amount of congestion in the casting bay, so that it is difficult or impossible to transfer all the ingots in the minimum track time. This furnace is at the end of the melting shop, and it is only possible to use two of the cranes, Nos. 3 and 4, when tapping. In consequence, two full ladles can be drawn off in rapid succession, but if the residual steel is to be tapped at the earliest time, these ladles must be transferred to cranes Nos. 1 and 2 for teeming, in order that Nos. 3 and 4 can pick up empty ladles and return to the furnace. If, on the other hand, either or both of cranes Nos. 1 and 2 are in use elsewhere in the casting bay, it is impossible for the residual steel to be tapped until the first two ladles have been teemed. Observations made on one cast, from tapping until the ingots were charged to the pits, are given in Table VIII, and shown graphically in Fig. 8. All four ladle cranes were used and their movements during the tapping and teeming periods are shown approximately in Fig. 9, indicating that because of the transfer of ladles some delays inevitably occur, although not necessarily of the magnitude observed.

Stripping again followed the order of teeming and, with the exception of one 'sticker,' track times of the ingots cast from the first ladle were less than 1 hr. Ingots from the second ladle could not, however, be stripped until the cranes had completed the first cast, and, similarly, the third cast could not be stripped until after completion of the second, and in consequence their track times were successively longer. The slight interruption in the sequence towards the end of the second cast, which necessarily caused a delay in the whole of the third cast, was entailed by charging hot metal to the mixer. Had the third ladle been tapped immediately after the second, the track time of these ingots would have been still longer by 14 min., but it would have been possible to begin the fettling and recharging of the furnace in a correspondingly shorter time.

Two or more of the fixed furnaces tapping within a short interval would, similarly, be expected to give rise to a measure of congestion in the casting bays and delays in track times of one of the casts. In addition, stripping could not be carried out if teeming were in progress at a station nearer to the soaking-bay junction.

Table VIII

OBSERVATIONS ON CASTS FROM TILTING FURNACE

Furnace Tapped at 3.21 p.m. (zero time), 3.26, 3.45, and 3.54 p.m.

Mould No.	Finish Teem	Strip		Charge to Soaker		Time in Mould		Track Time	
	Min.	Hr.	Min.	Hr.	Min.	Hr.	Min.	Hr.	Min
1	15	1	9	1	14		54		59
2	15	1	29	1	31½	1	14	1	16½
3	18	1	13½	1	15		55½		57
4	18	1	15	1	17		57		59
5	21½	1	16½	1	19½		55		58
6	21½	1	18	1	20½		56½		59
7	24½	1	19	1	22		54½		57½
8	24½	1	21½	1	23½		57		59
9	27½	1	23½	1	25		56		57½
10	27½	1	25	1	26		57½		58½
11	30	1	26	1	27½		56		57½
12	30	1	27½	1	30		57½	1	0
13	33½	1	31	1	34½		57½	1	1
14	35	Stub							
1	18	1	34	1	39	1	16	1	21
2	18	1	37½	1	41½	1	19½	1	23½
3	21	1	40	1	42½	1	19	1	21½
4	21	1	42	1	44	1	21	1	23
5	24½	1	43½	1	45½	1	19	1	21
6	24½	1	45	1	47½	1	20½	1	23
7	27	1	46½	1	50	1	19½	1	23
8	27	1	48½	1	52	1	21½	1	25
9	30	1	51½	1	54½	1	21½	1	24½
10	30	1	54	1	57	1	24	1	27
12*	33½	1	56	2	10½	1	22½	1	37
13	33½	1	58½	2	13	1	25	1	39½
14	37	2	14	2	19½	1	37	1	42½
15	37	2	15½	2	21	1	38½	1	44
1	33½	2	24	2	27	1	50½	1	53½
2	33½	2	25½	2	28½	1	52	1	55
3	38	2	27	2	30½	1	49	1	52½
4	38	2	29½	2	33	1	51½	1	55
5	41	2	32	2	35	1	51	1	54
6	41	2	35	2	36½	1	54	1	55½
7	43½	2	36	2	38½	1	52½	1	55
8	43½	2	38½	2	41½	1	55	1	58
9	46½	2	41	2	44	1	54½	1	57½
10	46½	2	42½	2	57	1	56	2	10½
11	50	2	44½	3	3	1	54½	2	13
1	44	Stub							

* Mould No. 11 was left empty

ANALYSIS OF RECORDS KEPT BY WORKS *A*

The time study at Works *A* comprises observations made on the individual ingots of two casts, and in

order to obtain further data the routine records were consulted. From these records the histories of all casts during a period have been plotted on chronographic charts. These charts are discussed on p. 81. Since the data are concentrated, an attempt has been made to examine several separate, but interrelated, aspects of the problem, starting with irregularities in open-hearth furnace operation.

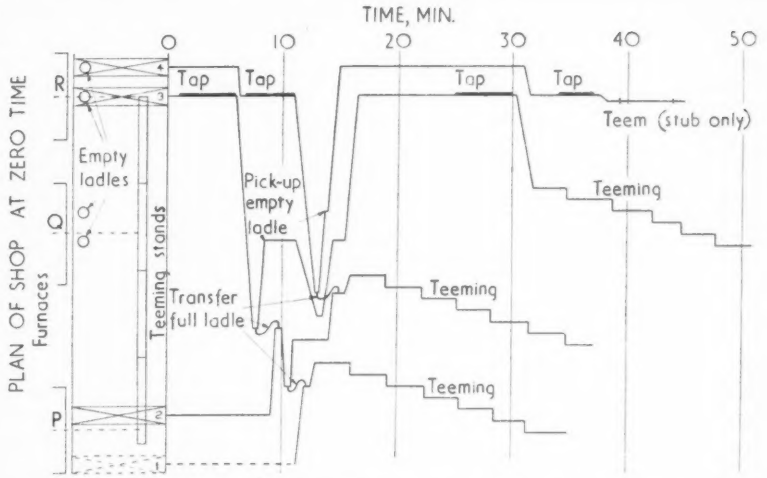

Fig. 9—Approximate crane movements during tapping and teeming of tilting furnace, Works *C*

Furnace Tapping Intervals

The supply of hot ingots to the soaking pits depends entirely upon the tapping intervals of the steelmaking furnaces. The average time to reduce an ingot to slabs at Works *A* is 3½ min., and each cast consists of 9–12 ingots, so that if two or more furnaces are tapped within any one period of about 30–40 min., only one of the casts can be heated and delivered to the cogging mill without delay. Figure 10 illustrates the irregularities in tapping intervals during the week 18th to 24th January, 1948, when the detailed observations were made. A charge was tapped from each furnace during Sunday night, the majority of these taps occurring between 10 and 12 P.M., and ' bunching ' occurred until, after three or four heats, the congestion was reduced by the natural variations in tap-to-tap times and by the inability to charge all the furnaces simultaneously.

The track times of the ingots of each cast are also shown in Fig. 10, and these times show a wide spread. The longest track times correspond to periods of intense activity in the casting bay, and the three batches of ingots which were allowed to go cold were cast during such periods. Since a cast may be rolled in 30–40 min., ' bunching ' cannot account for all the wide spread in track times, and one cause may be the lack of empty soaking-pit space at times when it is needed.

Tapping intervals are plotted in Figs. 11–13 as histograms, showing the frequencies of taps over ½-hour intervals. Figure 11 is taken from all heats in one furnace during 1944, and shows that as many as 10% of the tap-to-tap times fall outside each of the wide limits of 12 and 16 hr. Figure 12 is a histogram for the furnaces of both melting shops, and shows similar features to Fig. 11. In Fig. 13 the tapping intervals for the week 18th–24th January are compared with those for a somewhat longer period. A sample from so short a period as one week may not be completely typical ; in fact a high proportion of the taps are ' early,' and the peak of the histogram is less sharply defined than usual. The mean tapping intervals during the week considered and during a previous fortnight, are given in Table IX ; the exceptionally long heats at weekends have been excluded.

Prediction of the next tap of any furnace on the basis of the time of the last and of the expected tap-

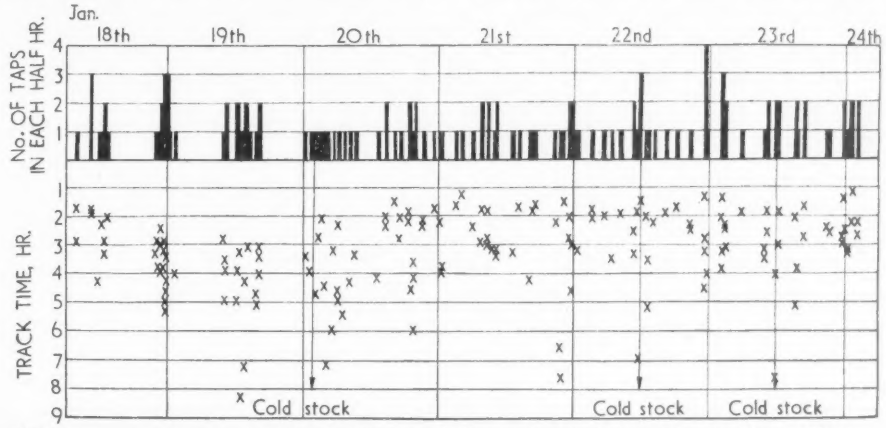

Fig. 10—Furnace taps and corresponding track times showing ' bunching.' From 6.0 a.m., 18/1/48, to 6.0 a.m., 24/1/48, Works *A*

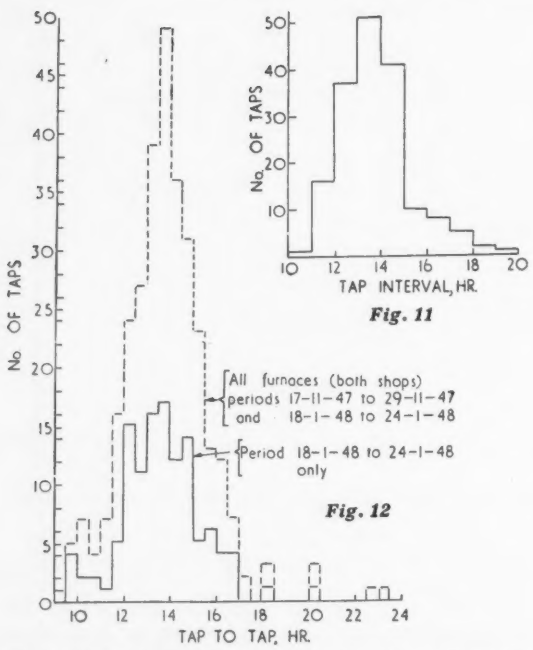
No. OF TAPS
TAP INTERVAL, HR.
All furnaces (both shops) periods 17-11-47 to 29-11-47 and 18-1-48 to 24-1-48
Period 18-1-48 to 24-1-48 only
TAP TO TAP, HR.

Fig. 11

Fig. 12

Fig. 11—Tap-to-tap time histogram of one furnace, Works *A*

Fig. 12—Tap-to-tap time histograms of all furnaces, Works *A*

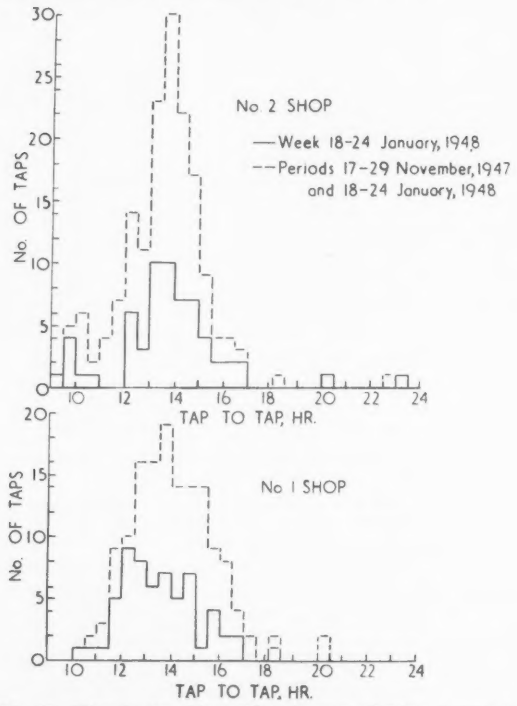
No. 2 SHOP
—Week 18-24 January, 1948
--Periods 17-29 November, 1947 and 18-24 January, 1948
No. OF TAPS
TAP TO TAP, HR.
No. 1 SHOP
No. OF TAPS
TAP TO TAP, HR.

Fig. 13—Tap-to-tap time histograms of Nos. 1 and 2 melting shops, Works *A*

to-tap times does not promise any useful accuracy because of the wide variations revealed by the standard deviations. It may be possible to predict with moderate accuracy the number of casts which will occur during a period of a few hours, and this may assist in deciding whether the pits will need to be charged with cold ingots to preserve continuity of rolling.

Track Times

Histograms have been prepared (Fig. 14) showing the number of ingots having various track times, during the week 18th to 24th January, 1948.

These histograms are summarized in Table X, which shows that ingots from No. 1 shop are delayed much more than those from No. 2 shop. The minimum

Table IX

TAP-TO-TAP TIMES

Shop	Mean: Week Considered, Hr.	Mean: Earlier Fortnight, Hr.	Standard Deviation: Week Considered, Hr.	Standard Deviation: Earlier Fortnight, Hr.
No. 1	13·63	14·59	1·59	1·63
No. 2	13·80	13·97	2·60	1·67

Table X

TRACK TIMES

Shop	No. of Ingots	Track Time, hr.: Minimum	Track Time, hr.: Mean	Proportion less than: 3 hr., %	Proportion less than: 2 hr., %
No. 1	578	1·75	3·62	20	4
No. 2	637	1·2*	2·45	66	49
Total	1215		3·00		

* Two ingots with times slightly less than 1 hr. have been ignored.

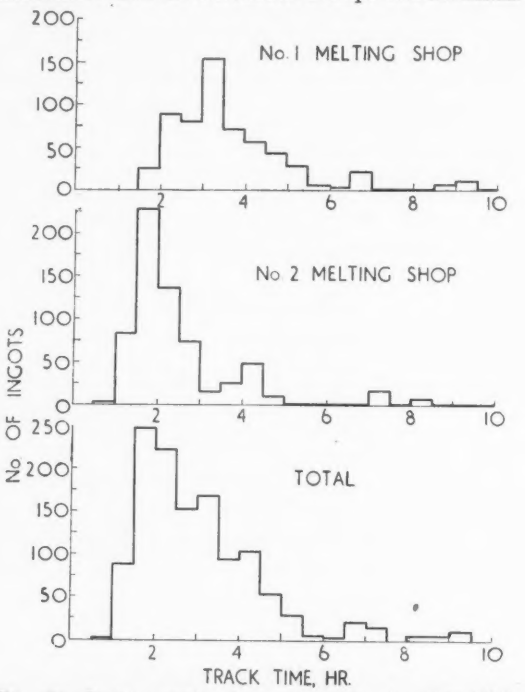

Fig. 14—Ingot track times, Works *A*, 18th–24th January, 1948

times given show that it is possible for the extra distance travelled by ingots from No. 1 shop to account for only 0·55 hr. difference in track times between the two shops. The difference is often much greater, however, and between the mean track times it is 1·17 hr. Some of the delay may be due to traffic hold-ups *en route* from No. 1 shop, but a further cause is a possible lack of liaison between the soaking pits and the No. 1 shop. Table XI gives examples of occasions when, although casts had already been made in No. 1 shop, cold ingots were charged to the soaking pits and the hot steel was delayed because no soaking pit was then available.

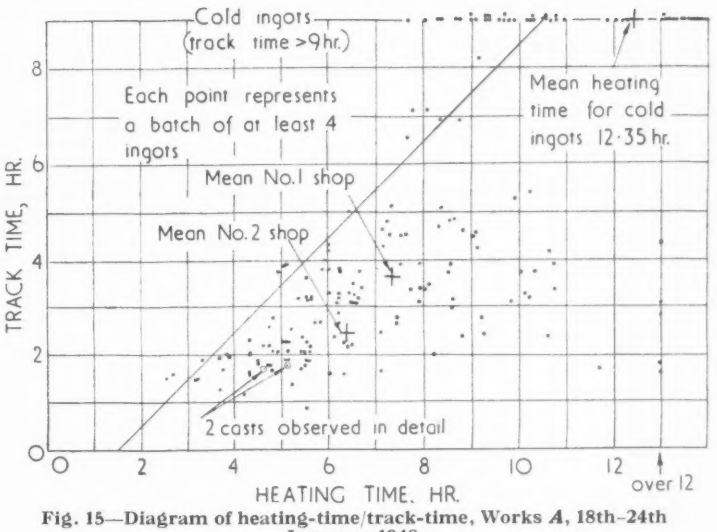

Fig. 15—Diagram of heating-time/track-time, Works *A*, 18th–24th January, 1948

Heating Times

The main factor which should influence the length of time each ingot remains in a soaking pit is the track time. Figure 15 is a plot of heating times against track times, each point representing a batch of ingots. It is not always possible to charge the whole of one cast to one pit, so that each cast may give rise to one or more points in Fig. 15. No definite relationship can be deduced, which implies that heating times are over-influenced by other factors, and these may include an intentional factor of safety to ensure that soaking is complete. In addition, when several pits are holding fully soaked ingots at one time, some are delayed until the mill is ready to receive them.

In Fig. 15 all the points for hot and warm ingots (*i.e.*, with track times less than 9 hr.) fall to the right of the line :

$$\text{Heating time} = \text{Track time},$$

but it appears that ingots are ready for rolling if soaked for times given by the relation :

$$\text{Heating time} = \text{Track time} + 1\tfrac{1}{2} \text{ hr.},$$

there being sufficient examples of such a standard of performance. Further, it is possible to heat cold ingots in 10 hr. Thus, even if track times could not be reduced, some saving in heating times appears to be possible. In order to assess the magnitude of these savings some further information is needed, and this is given in Table XII.

The cogging mill has a utilization factor of nearly 70%, and during the time it is operating the average rolling time per ingot is 3·48 min. This compares favourably with the 3 min. which is the minimum time noted on even short periods. The tonnage rolled during this week (11,000 tons) is typical of the output of Works *A* at the time of the study.

Table XII shows that, on the average, heating times are at least 2 hr. longer than those given by :

$$\text{Heating time} = \text{Track time} + 1\tfrac{1}{2} \text{ hr., or 10 hr. (whichever is shorter)}$$

so that a saving of 25% in the pit usage is to be aimed at, even with present track times. This means either that 25% of the pit capacity could be off gas at any one time, or that a further 300 imported ingots could be heated from cold, increasing the mill output by 18%. In order to confirm this estimate and to show how the savings could be effected, a complete programme of soaking-pit operation has been worked out. This programme is discussed below.

Table XI

DELAYS TO INGOTS FROM NO. 1 SHOP

Pit No.	No. of Cold Ingots Charged	Casts* Already Teemed, With Their Subsequent Track Times, hr.	
5	13	C†	4·5
		D	3
		E	3·9
12	6	E	4·5
3	12	K	4·9
10	6		
7	13	E	3·3
9	6		
2	12	E	4·3
5	13	D	4·0
		F	6·9
10	6	J	3·5
11	6		
12	6		
3	12	E	3·5

* An average cast consists of 9 ingots.
† Furnace identification letter.

EFFECT OF IMPROVED OPERATION

It has been suggested already that there is a lack of liaison between the soaking pits and the melting shops, and a failure to draw the heated ingots when they are ready for rolling and the mill is ready to receive them. It is not possible to see the effects of improvements in these respects by examining the time studies already given, since small improvements only become effective when they have accumulated

Table XII

COGGING MILL AND SOAKING PIT THROUGHPUTS—WEEK, 18th TO 24th JANUARY, 1948—WORKS *A*

Source of Ingots	No. of Ingots	Track Time, hr.		Mean Heating Time, hr.	Minimum Heating Time, hr.		Pit Usage, %
		Mean	Minimum		Based on Mean Track Time	Based on Minimum Track Time	
No. 1 shop	578	3·62	1·75	7·28	5·12	3·25	31·5
No. 2 shop	637	2·45	1·2	6·35	3·95	2·7	30·3
Cold stock	117						
Cold ingots cast during week considered	63			12·35	10·0	10·0	38·2
Imported ingots	233						
Total ...	1628 (= 11,000 tons)			8·21			100

Mill in operation 94·3 hr. (in 40 separate spells) during 136-hr week, *i.e.*, 69·3% usage.
Average rolling time 3·48 min. per ingot.

over a long period. Accordingly, a completely revised programme of charging, drawing, and rolling has been worked out, based on the actual tapping times of the steelmaking furnaces in the week 18th–24th January, 1948. A series of working assumptions, based on the time studies, has been made. The effect of these has been to reduce track times and heating times towards limits which have already been shown to be attainable in practice.

The revised programme was worked out on a chronographic chart similar to that given in Fig. 18. The application of such charts is discussed on p. 81.

Assumptions

The following working assumptions are based on observations already discussed. Reference to these observations is made by the page numbers in parentheses.

Table XIII

HYPOTHETICAL THROUGHPUT

Shift	Ingots Rolled	Cold Ingots Charged
1	117	18
2	122	72
3	103	28
4	99	50
5	117	24
6	115	40
7	113	50
8	113	46
9	100	19
10	117	51
11	120	40
12	113	54
13	114	24
14	123	45
15	114	34
16	111	12
17	113	0
Total	1924	607

The total of 1924 is made up of:
Hot ingots from No. 1 shop ... 582
Hot ingots from No. 2 shop ... 605
Cold ingots from stock 607
Ingots initially in pits 130
} 1924

(i) There is no attempt to forecast furnace tapping, but the heater knows of any cast as soon as teeming is completed

(ii) The minimum track times possible are: No. 1 shop, 2 hr.; No. 2 shop, 1½ hr. (p. 78)

(iii) The minimum heating time allowed is equal to 1½ hr. plus the track time, or to 10 hr. in the case of cold ingots (p. 79)

(iv) The mill is capable of rolling one ingot every 3½ min. (p. 79), beginning 10 min. after the start of each shift and continuing until 10 min. before the end, with a half-hour break in the middle of each shift. This means that the mill has a maximum capacity of 123 ingots per shift, or 2091 ingots in a 17-shift week

(v) Each pit requires cleaning, and the addition of a new bottom of breeze once in the middle of the week, a further cleaning being possible during the weekend. Cleaning and subsequent reheating of each pit occupy 1½ hr

(vi) Pit capacities are: Nos. 1–4, 14 ingots each; Nos. 5–8, 13 ingots each; Nos. 9–12, 6 ingots each

(vii) At the start, each pit is holding the same number of heated ingots as at 10 p.m. on Sunday, 18th January

(viii) Since works *A* imports cold ingots as well as rolling the cold ingots from, for example, weekend working, the supply of cold ingots is sufficient to meet any increased demands

(ix) Except at periods of pit cleaning, only one handling crane need be used in the soaking-pit bay. This gives ample allowance for possible breakdowns

(x) In neither shop can two casts be stripped simultaneously; stripping of one cast takes about 15 min. (p. 68).

These assumptions do not include any which are not justified by examples of practice at Works *A*, except that liaison between the soaking pits and the melting shops is assumed. With this liaison, a pit should be reserved for each cast so that its track time is the minimum whenever possible. In general, this means that cold ingots are not charged to a pit which becomes empty 1½ hr. after a cast in No. 1 shop, or 1 hr. after a cast in No. 2 shop.

Results

Using these assumptions, the throughput achieved was 1924 ingots, made up as shown in Table XIII. The actual throughput during the week in question was 1628 ingots, so that the increase is 18·2%. In other words, the utilization of the mill has been

Table XIV

MEAN HEATING TIMES

Ingots	Hypothetical Hr.	Actual Hr.
From No. 1 shop	5·74	7·28
From No. 2 shop	5·17	6·35
Cold ingots	11·46	12·35

Table XV

MEAN TRACK TIMES

Shop	Hypothetical Hr.	Actual Hr.
No. 1	2·71	3·62
No. 2	2·09	2·45

increased from 78% to 92·2% of the assumed maximum capacity of 2091 ingots per week.

The immediate cause of the increase in throughput is a reduction in the mean heating times. These are compared in Table XIV with the actual mean heating times for the week 18th to 24th January, 1948.

In few cases was the mill free to accept ingots after the minimum heating time; this is shown pictorially in Fig. 16, which is a scatter diagram of heating time against track time, and may be compared with Fig. 15. The improvement in track times is given in Table XV.

A high proportion of the track times (68% in No. 1 shop, and 62% in No. 2 shop) is within half an hour of the minimum allowed. Track time histograms are given in Fig. 17 and may be compared with Fig. 14.

Pits 10, 11, and 12 were always charged with cold ingots, as was pit 9 in all but three cases. The number of cold ingots charged to these four pits was 264 (282 including the three batches of hot ingots). Thus, without these four pits (or their equivalent in capacity) the total throughput would have been at least 1642 ingots. This exceeds the present weekly throughput.

An independent calculation shows that, had *all* the ingots been charged to the soaking pits in the cold condition, the week's throughput (on the same assumptions as before) could have been 1452 ingots; 89% of the actual throughput when 75% of the ingots were initially hot.

GRAPHICAL RECORD KEEPING

It has been shown that the throughput of the soaking pits might have been improved considerably by using all the knowledge available, by planning the charging and drawing programmes to reduce delays in track times, and by avoiding very long heating times. No attempt is made to forecast events, and no information is required which is not already recorded on the soaking-pit record sheets, but the value of this information is not clear until it is presented graphically. It is difficult to visualize the position at any time from a number of record sheets, but it is possible to do so with a chart of the type used in this investigation. Long track and heating times are immediately obvious, and hastening action can be taken before the delay becomes excessive.

Explanation of Chart used in Investigation

An example of the chart is given in Fig. 18. Figure 19 is an extract from such a chart, sufficient to show the meanings of the various symbols adopted. Each steelmaking furnace is represented by a horizontal line in the top half of the diagram, and each soaking

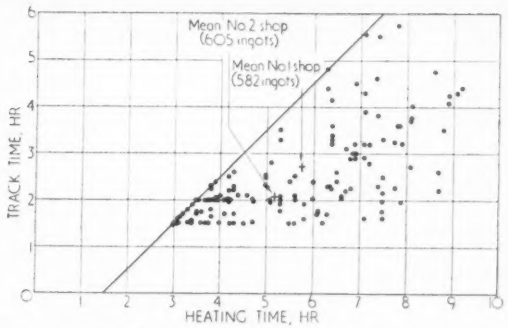

Fig. 16—Heating-time/track-time for hypothetical operation, Works *A*

pit by a horizontal line in the lower half. Further lines represent cold stock and the mill. The horizontal axis is a time scale (1 in. representing 1 hr.). As each cast is teemed, a mark is made on the furnace line at the appropriate time, and the cast number, number of ingots, and any other details are added. Charging to the soaking pits is represented by a vertical line connecting the furnace line to the appropriate pit line, and a symbol represents time spent in charging. At each movement, the number of ingots involved is added to the diagram. Since heating must occupy a certain minimum time, a point can be plotted giving the time when each batch of ingots should be ready to be drawn. When drawing actually takes place symbols are placed both on the soaking pit line and on the mill line, showing the time taken and the number of ingots rolled.

Using these symbols, the diagram can be made sufficiently clear in black and white, but it may be

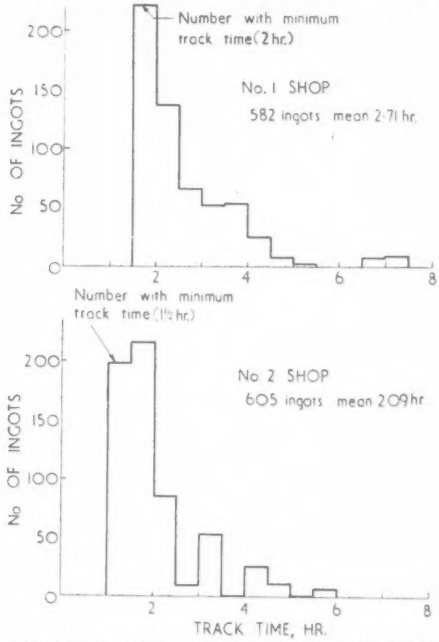

Fig. 17—Track time histograms for hypothetical week's working in Nos. 1 and 2 shops

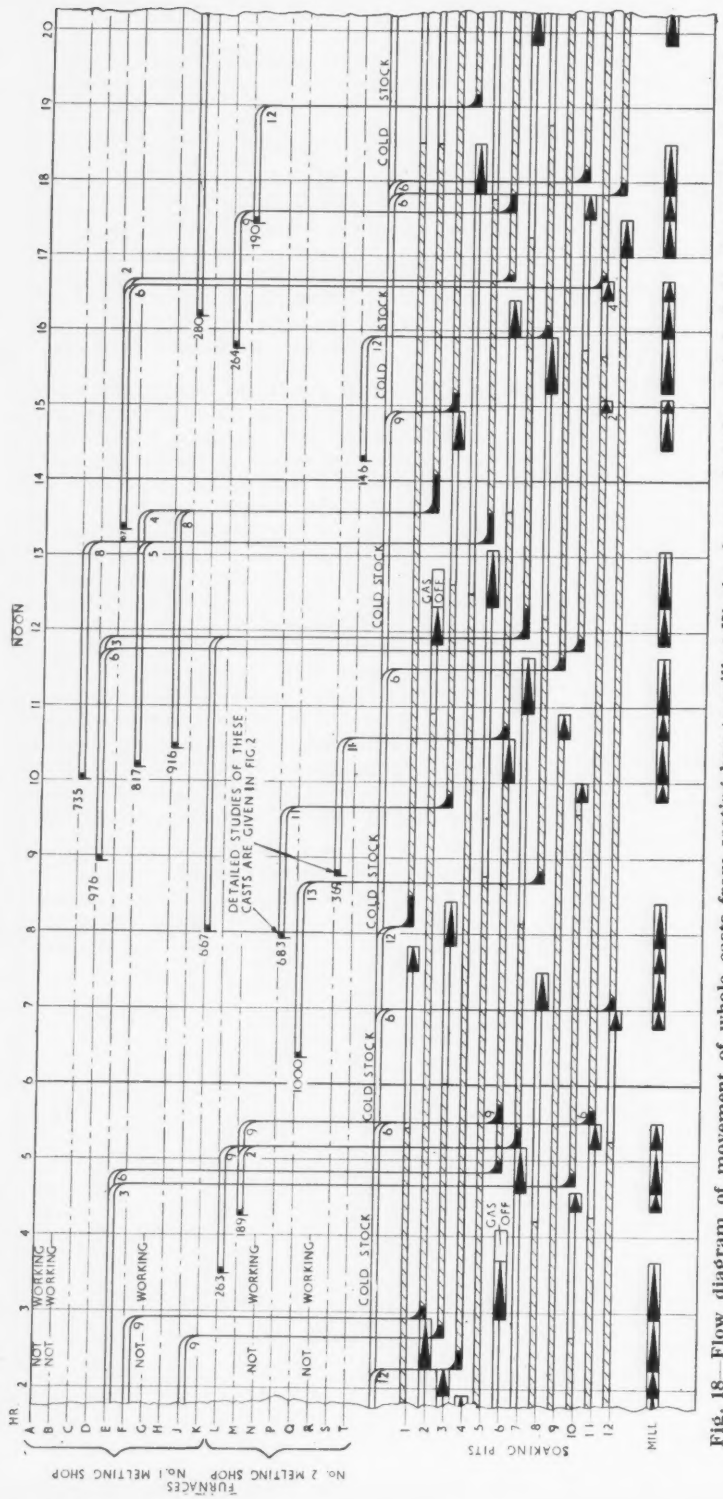

Fig. 18—Flow diagram of movement of whole casts from casting bay to mill at Works *A*, covering the period of the time study in Fig. 2. For explanation of symbols, see Fig. 19

an advantage to use colours. It may assist forward planning if expected events are indicated temporarily on the chart by means of pins, and replaced by a permanent record when the events actually occur. There is ample space on the diagram for notes about the special characteristics of each cast and for notes of crane breakdowns and other delays, so that the present system of recording might be dispensed with.

DISCUSSION

As a soaking pit is a batch furnace from which the combustion gases leave at a high temperature, a great improvement in fuel consumption per ton of steel may be obtained by reducing the time the stock remains in the furnace. There are four possible ways of effecting this reduction :

(i) By increasing the rate of heat transfer to the stock by a change in pit design

(ii) By defining, more accurately than is at present possible, the exact requirements for heating ingots to a fit state for rolling. This requires the determination of internal temperature distributions in ingots at all stages of cooling and heating. It may then be possible to reduce with confidence the factors of safety at present allowed

(iii) By insulation, to reduce the heat lost between stripping and charging

(iv) By reduction in track time to reduce heat losses.

Examination of the first three possibilities requires a knowledge of temperature distributions within ingots. The necessary experimental techniques for obtaining this knowledge, which are in course of development, are discussed on p. 84. This paper is an examination of the fourth possibility, that of reducing mean heating times by reducing mean track times.

Reduction of Track Times

Track time divides naturally into two parts : (i) from teeming to stripping ; and (ii) from stripping to charging. The minimum value of the first part is determined by the need to form an adequate solid shell before mechanical handling can be undertaken with safety. An

upper limit to this minimum time, sufficient for the purpose of this paper, has been found by observing the times allowed in practice at three works. More accurate knowledge is desirable.

The time between stripping and charging to the pits is determined by the mechanical handling arrangements and the availability of space in the soaking pits. Variations in these times were found within each of the three works visited. Not only were there variations between casts but also between the individual ingots of each cast. The causes of the variations in these times include :

(i) The orders in which ingots are handled during teeming, stripping, and charging to the pits causing variations within each cast

(ii) Lack of soaking-pit space when it is needed causing differences between casts

(iii) Inability to conduct two casts simultaneously through the same stage of the process, and in some cases through two different stages of the process, causing differences between casts.

These causes are further discussed below.

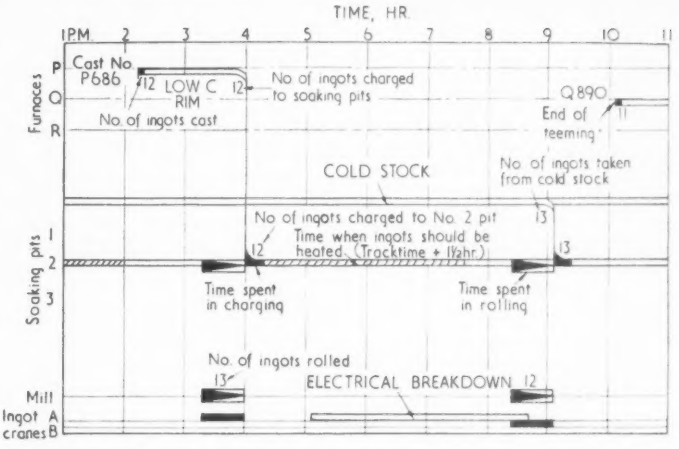

Fig. 19—Explanation of symbols used in chronographic chart, Fig. 18

Effect of Works Layout

Order of handling is dictated by the works layout and the mechanical handling arrangements. For example, the practice of stripping ingots in the reverse order of teeming is caused at Works *A* by the poor visibility from the stripper crane cab, and at Works *B* by the layout of the railway track between the melting shop and the stripping bay. At Works *C*, the more logical procedure of stripping and teeming in the same order is followed, because it is convenient.

Where the casting and soaking-pit bays are not in the same building, ingots must, at some stage, be transferred by rail, and there are two principal methods. In the bogie casting method, ingots are hauled in their moulds by locomotive and then stripped near the soaking pits, whereas with pit casting, ingots are first stripped and then transferred by rail.

Diamond[3] has criticized bogie casting from the points of view of maintenance and locomotive efficiency but, from the fuel economy point of view, this method has advantages. It is preferable that stripping should immediately precede charging to the soaking pit, since heat is lost at a greater rate from a stripped ingot than from one still in the mould. The bogie-casting method, with the soaking pit and stripping bays under the same roof, allows stripping to be under the jurisdiction of the soaking-pit heater. With pit casting, however, ingots are delayed between stripping and charging by a time depending on the availability of a locomotive and of space in the soaking pits. Further, since with bogie casting the essential journey occurs during the solidification period, which must be allowed in any case, bogie-cast ingots can have shorter overall track times than pit-cast ingots.

With the casting, stripping and soaking-pit bays virtually in the same building, handling by crane offers the maximum flexibility and speed of handling. For example, at Works *A* the ingot transfer car is not loaded until stripping is completed, whereas at Works *C* each ingot is moved by crane to the soaking pits as soon as it is stripped.

Availability of Soaking-Pit Space

Differences between the track times of whole casts are due partly to lack of soaking-pit space when it is needed. To some extent, this is because the soaking pits absorb the output of a number of steelmaking furnaces tapping at irregular intervals, and feed a single cogging mill capable of rolling at a steady rate. Poor liaison between the melting shop and soaking-pit operators, who often have little warning of the arrival of a cast, is an aggravating factor.

Lack of soaking-pit space is largely caused by long heating times, which are due partly to the long track times caused by lack of soaking-pit space. Thus, there is a vicious spiral, so that any reduction of track or heating times is capable of causing a substantial improvement in soaking-pit throughput, which will be reflected in an increase of soaking-pit efficiency.

Magnitude of Possible Savings

Having shown, as a result of a critical examination of the time studies, the possibility of reducing track times by systematic handling, it was necessary to find the effect of this possibility on the vicious spiral already mentioned. This could only be done by estimating the effect over a long period, and a whole week's soaking-pit programme was worked out for comparison with an actual week's working. In this hypothetical example, the operation of the steelmaking furnaces was the same as that recorded, but it was assumed that ingots could always be charged after the minimum track time if a soaking pit were available.

Comparison between the hypothetical and the actual operations showed that the soaking-pit throughput might be increased by 18%, the additional ingots required being drawn from the cold stock. A greater increase would be possible if more hot ingots were available. Alternatively, the soaking-pit capacity might be reduced by 18% without detriment to the

present throughput. Either of these results gives an increase in the overall efficiency of the process at least as great as any to be expected from increase of the thermal efficiency of the soaking pit itself.

Throughout the investigation, graphical methods of representation were found invaluable. It is suggested that if records of the histories of all casts were kept in this form, an important advance would be made on the present methods, employing log books or record sheets. Thus, it would be possible to plan the soaking-pit programme with greater accuracy, and to ascertain at a glance the positions of the ingots and the durations of the various handling operations.

Moreover, such a graphical method of recording data would enable delays to be appreciated in their early stages, so that remedial action could be taken. By thus reducing delays, the throughput per pit would be increased, and fuel would be saved.

Future Developments

The recommendations suggested for improving performance have not yet been put to the test, and it is hoped that some works will examine the possibilities. In the first instance, assessment must necessarily be by trial and error, and will depend on different opinions of suitable temperatures for rolling steel, not only between one works and another but also within one works. It is fully realized that one roller may reject steel which another would accept for the mill, and at present there is no criterion available.

A knowledge of temperature distributions at all stages of cooling and reheating would enable the process to be put on a sound scientific basis. In general, the temperature of the molten steel can be measured immediately before tapping, and by modifying the technique of using immersion thermocouples it should not be unduly difficult to measure the cooling which occurs over the period of 20 to 30 min. during which the individual ingots are teemed. Existing techniques do not, however, permit the measurement of temperature distributions during the successive stages.

The rate at which heat is transferred to the mould by radiation and conduction is largely a matter of conjecture. The unknown and varying heat-transfer coefficients between ingot, mould, and surroundings, together with the varying diffusivity and conductivity of the steel, and the need to consider the latent heat of solidification make it impracticable to determine analytically the temperature distribution at any stage. It is hoped, however, that by measurement of the actual surface temperature as a function of time, further information can be obtained, either by calculation or by the use of such devices as the differential analyser.

During the past year, the Physics Department of the British Iron and Steel Research Association has been developing surface pyrometers which will enable temperatures of moulds and ingots to be measured under casting-bay conditions. The work carried out suggests that the difficulties so far encountered are not insuperable, and that prototype instruments may be available in the near future. This work is to be extended to cover the measurement of the surface temperature of steel in the soaking pits.

With these instruments and technique, it will be possible to ascertain the temperature distribution and heat content at all stages. This knowledge will throw light on the following problems :

Solidification—Determination of the time at which the shell of the ingot has strength to withstand, without deformation, the ferrostatic head of liquid core. This is the earliest time at which the ingot can be stripped, but further cooling may be necessary to prevent deformation on reheating.

Drawing—Determination of the time at which the ingot has any desired temperature distribution.

Thermal Efficiency—At some stage of cooling, the heat content of an ingot is equal to that required for rolling, and if further heat loss could be prevented the ingot would, in time, attain a uniform temperature throughout. This, of course, is the principle of the original Gjers pit. Since ingots could be prepared for rolling in an insulating container it follows that the thermal efficiency of a fuel-fired pit may be low, or even negative. Knowledge of the heat contents of hot charged ingots will make it possible to measure efficiency during furnace trials.

Pit Design—Further, such furnace trials, under a variety of controlled conditions will make it possible to determine the significance of the various aspects of pit design, which will enable ingots, stripped at the optimum time, to be prepared for rolling with minimum fuel consumption.

Ingot Storage—The time studies have suggested that, with open-hearth furnace operation, it will seldom be possible to process more than 60% of the ingots under such optimum conditions. This raises the question of whether there is any saving possible by storing the remaining 40% in 'dead' soakers, with little loss of heat, before charging to a live pit. Such a procedure would also reduce the scaling losses caused by prolonged holding at a high temperature.

The fields in which further work is required are :

(i) Practical study of improvements in performance arising from closer control of operating sequences

(ii) Development of surface pyrometers (*a*) for ingot moulds at temperatures up to 700° C. ; (*b*) for stripped ingots at temperatures up to about 1100° C. ; (*c*) for ingots in soaking pits at temperatures up to about 1350° C.

(iii) Development of a technique for establishing the temperature distribution through an ingot at any time

(iv) Study of the rates of cooling of stripped ingots in the open and in insulated chambers, and of the rates of reheating to rolling temperatures with various furnace conditions

(v) Study of the influence of soaking-pit design on heat-transfer coefficients and thermal efficiency.

Acknowledgments

The authors express their thanks to the Companies at whose works the time studies were made and whose staffs gave every assistance.

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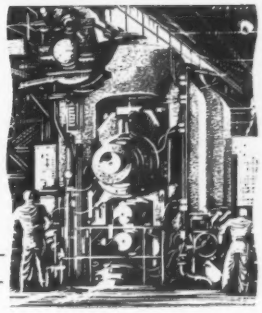
IRON AND STEELWORKS ENGINEERING



THE IRON AND STEEL ENGINEERS GROUP

THE IRON AND STEEL ENGINEERS GROUP

REPORT OF THE TENTH MEETING

THE TENTH MEETING OF THE IRON AND STEEL ENGINEERS GROUP of The Iron and Steel Institute was held at 4 Grosvenor Gardens, London, S.W.1, on Wednesday, 22nd June, 1949. Mr. W. F. CARTWRIGHT (Steel Company of Wales, Ltd.), Chairman of the Group, presided. The Discussion on "Continuous Steel Strip Pickling" by R. W. Treasure was published in the December, 1949, issue of the *Journal*.

Discussion on—

THE DISPOSAL OF SPENT ACID PICKLING LIQUOR*

Mr. **W. B. Wragge** (Exors. of James Mills, Ltd.) gave a summary of his paper and illustrated it with slides.

Dr. **J. Pearson** (British Iron and Steel Research Association): Because of the difficulty of finding markets for copperas or ferrous sulphate monohydrate, the problem of disposing of any products of treatment may be said to reduce itself to one of finding a process in which no waste is produced and of which the products of treatment are usable inside the steelworks.

The Spent Pickle Liquor Sub-Committee of the British Iron and Steel Research Association, set up to look into possible methods of tackling the disposal problem, has concluded that the only feasible way is to convert the spent pickle liquor to iron oxide and sulphuric acid. This would, of course, involve the separation of ferrous sulphate, the return of the mother liquor to the pickling tanks, and the conversion of the ferrous sulphate into sulphuric acid. Those members of the Sub-Committee who were acid makers concluded that a process involving roasting and contact catalysis could make sulphuric acid at a cost comparable with that at which it could be bought only if it were worked on a scale far larger than anything we could contemplate. It was thought that there might be a method whereby the acid could be made as a dilute acid of a strength suitable for pickling, and which would be economic even on a smaller scale.

A process which promises to fulfil these conditions has been described by Keyes† in America. It depends upon roasting ferrous sulphate and submitting the gases to an autoxidation process. Keyes' scheme, of which our own is a modification, consists of taking part of the ferrous sulphate which is recovered, dissolving it in water and using that to fill a tower through which are passed, along with air, the gases from the roasting of the rest of the ferrous sulphate. A combination of sulphur dioxide, oxygen, and water takes place in the tower and the acid is then run back into the pickling tanks. As far as I am aware, the Keyes' process is theoretical only, but the conclusions reached after a lengthy study of it have been quite promising.

The process as we envisage it involves separation of the ferrous sulphate. During the past year we have determined the minimum temperature necessary to get effective decomposition of the ferrous sulphate, and the best things to add to speed it up. We have also studied the conditions under which the gas, which is mainly sulphur dioxide, can be converted to sulphuric acid. We can now produce in the laboratory, from gases which are typical of those which come from a roasting plant, sulphuric acid of a strength which would be adequate for any pickling operation. Sulphur dioxide conversion efficiency is 99·9 to 100%. We can very easily make acid for normal pickling, where conditions are not very stringent, and by taking suitable precautions we can make acid strong enough for the most drastic requirements.

The Sub-Committee has authorized the design of a pilot plant in which we hope to make sulphuric acid of the order of one ton per day in the form of a dilute solution. The design of the pilot plant is at present being studied by firms of chemical engineers.

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* *Journal of The Iron and Steel Institute*, 1949, vol. 62, June, p. 213.

† H. E. Keyes, *Chemical and Metallurgical Engineering*, 1946, May, p. 126.

Mr. **H. Edwards** (Messrs. John Summers and Sons, Ltd.): I recommend that when acid strength is referred to the relation weight per weight should be used, because reference to percentage, or percentage volume, gives no idea of the actual tank conditions.

There are two stages to the problem of disposal. Firstly, the recovery of waste acid—well-established processes are available for this, and newer processes have been investigated. Secondly, the disposal of the iron sulphate or copperas, there being an urgent need for an economic outlet for copperas, since a plant which is pickling 8000 tons of strip produces approximately 120 to 150 tons of copperas. It is necessary to produce oxides which are easily broken and easily soluble, and to wash the strip with water sprays before immersion in the acid, to prevent loose scale entering the acid tanks, where it is dissolved with the formation of more iron sulphate.

Economy in the use of acid is very important. Overflow is a problem, particularly with direct heating, and a certain amount of indirect heating is necessary to prevent a very serious loss of acid. The losses by evaporation should be reduced by the use of proper covers.

Mr. **N. Swindin** (Nordac, Ltd.): There are about ten ways of expressing the concentration of a pickling liquor. The ternary diagram saves an enormous amount of trouble, and gives a visual picture of what takes place, not only in pickling, but also in all recovery processes. It is worthy of serious study. The submerged combustion process of pickle recovery depends largely upon the solution of the problem of removing the deposited sulphate from the system. To make the process more acceptable to the steel industry, a lot of time was spent in recovering the heat from the exit gases and steam. It was found that if the exit gases are cooled to dewpoint, what is called a dry K is obtained, and with continued cooling a wet K results. This requires fairly high velocities, and additional pressure on the system. Three-fifths to four-fifths of the energy required to give these high velocites came from the burner and not from the blower, showing that some advantage was being taken of the increase in volume of the gases in passing through the burner.

Mr. **J. Bonelle** (Prodorite, Ltd.): The mother liquor reclaimed from acid recovery plant is put into the pickle bath and fresh acid is added. Has the author any information to confirm our experience that, so far, one cannot entirely eliminate the iron content of that mother liquor? With the crystallization of copperas that takes place, the form of lining to be used in the tanks is very important.

I should like to hear what the author has to say about hydrochloric acid effluent.

Mr. **W. B. Wragge**: From a consideration of the equilibrium diagrams it is obvious that it is impossible to eliminate all the iron in the liquor unless the temperature is reduced to many degrees sub-zero. For refrigeration down to about minus 4 or 5° C. we find that we can reduce the iron content from an initial 8% in the spent liquor to about 2 to 2½% in the recovered liquor, when bulked up to the original volume. The diagram indicates that the iron drops from 8 to about 3%, but this has to be diluted with water because the volume is less than the original volume. That is our normal range.

I deliberately did not deal with hydrochloric acid in any detail, because it appears that it is used to a very minor extent in the steel industry for descaling purposes. I understand that, in those plants where it is used, disposal to the chemical industry of the spent liquor, containing about 33% or more iron, is very satisfactory.

Mr. **J. H. G. Willan** (Nettlefold-Stenman Hinges, Ltd.): Can Dr. Pearson tell us the approximate quantity of copperas required for the production of one ton of dilute sulphuric acid? Should there be one central plant to which users would send acid, or is the plant to be of such a character that the small user could install it to meet his own particular needs?

Dr. **J. Pearson**: It is not known what, in practice, will be the overall efficiency of the plant. Obviously, there are certain losses of sulphur dioxide in the roasting. The efficiency with which the sulphur dioxide is converted to sulphuric acid will, it is hoped, be at least 99%. Assuming something of the order of 90% as the overall efficiency of the process, then one ton of H_2SO_4 as a dilute solution will require 3·1 tons of copperas or 1·9 tons of monohydrate.

We cannot say how small firms will be affected, since we do not know how the process will work. That is why we are having a pilot plant. We hope that the process will produce dilute acid economically even on a small scale, although I should say that the Sub-Committee is concentrating its attention in the first place on solving the problem for those steel firms which produce spent pickle in very large volume. If the process is not satisfactory for the smaller firms we shall turn our attention to finding some suitable way for them.

Mr. **Wragge**: I gather that the autoxidation process is designed to produce relatively dilute sulphuric acid. Have you any idea of the concentration?

Dr. **Pearson**: At the present time, in the best conditions, we can produce 50% acid, but we shall attempt in the first instance to make 25% acid. If the results are not better than that, we shall try modifications to remove contaminants which are responsible for keeping the acid concentration down to low figures. As far as the process itself is concerned, it will make acid stronger than 50%, but at concentrations over 50% the efficiency with which sulphur dioxide is removed from the gas is less than 100%. We can obtain up to 60% acid, but only with 80% efficiency.

Mr. **Wragge**: Has Dr. Pearson yet worked out a schematic flow sheet which will embrace the complete recovery of acid from spent liquor? In the plant that I am associated with, a constant tank volume is kept by the combined use of indirect and direct steam heating. By making additions of sulphuric acid in the dilute form, an increase in the volume of liquor in the pickle bath will be obtained, which means using more indirect steam heating and too much evaporation. If direct steam heating is used, the volume will increase and there will be overflow problems. With indirect heating only, there is thermal inefficiency. What balance can be created if weak acid is used for making bath concentrations? At the moment BOV is added because it contains only about 20% of water. I think there might be difficulties if recovered acid were added.

Dr. **Pearson**: The actual problem will depend upon the acid strength obtained. I cannot give any definite information about the water balance of the system. It may happen that with the dilute acids produced there will be an accumulation of water in the system; this will mean that some sort of concentration will have to be applied to acid made in the autoxidation tower.

Mr. **F. A. Batty** (Head Wrightson Machine Co., Ltd.): An efficient scale breaker loosens the scale very well, but creates an accumulation. An additional scale breaker, with an efficient means for removing the scale accumulation, would help to ease the burden on the pickle tanks, and I think that would be of some benefit. I consider that scale breakers in pickle lines have not been treated with the importance which they deserve.

Mr. **Wragge**: If strip, coated with scale, or the iron rod on which I have done my experiments, is bent so that the elastic limit is just about exceeded, most of the scale breaks off. Considering the degree of pressure required on sheet for the purpose of raising the surface to such stress that it reaches or just exceeds the elastic limit, would not the power requirements on the scale-breaker rolls be rather high? I think it is a good idea to remove the scale at the scale-breaking stage.

Mr. **Batty**: More power would be required to remove scale than for actual levelling, but the amount would be quite moderate and well within design limits.

The **Chairman**: Dr. Pearson said that the expense of making acid by the acid recovery process is greater than by making it in the normal way. On the other hand, the expense of getting rid of the acid is already high, and to that extent there is something on the credit side.

Mr. Wragge has presented the problem in such a way that the average engineer can clearly understand it.

Discussion on—

THE DESIGN AND OPERATION OF ANNEALING PLANT FOR MILD-STEEL SHEETS AND COILS*

Mr. **R. D. Pollard** (Messrs. Samuel Fox and Co., Ltd.): presented his paper.

Mr. **R. L. Willott** (Messrs. John Summers and Sons, Ltd.): Can the author tell me whether, at temperatures involved in normal annealing practice, there is any metallurgical effect in the outer skin of the material caused by oils remaining from the cold-rolling process?

Mr. **Pollard**: Most people keep to sub-critical annealing for mild steel, and this is particularly necessary for high-carbon steels. Either carburization or decarburization can be produced, according to the equilibrium of the atmosphere in the container, if a temperature of 700° C., or just over, is reached, but it is our experience that with temperatures of 670–680° C. usually adopted for sub-critical annealing, carburization or decarburization does not occur to any perceptible extent. From a structural standpoint, therefore, the presence of rolling-oil residues seems to have little effect under these conditions.

Mr. **O. J. Thomas** (Glynhir Tinplate Co., Ltd.): Has the author any experience of the different rates of cooling on mild steel, because when one compares annealing in batches and normalizing, there is quite a marked influence on the product?

The following results were obtained and published in my Presidential Address to the Swansea and District Metallurgical Society in 1944:

"The advantages of slow cooling below 650° C. are shown by the following test results, which were obtained by cutting a sheet in half and sending the two halves through a normalizing furnace at the same time, removing one half as it left the end of the cooling cycle, whilst the other half was stacked in the pile as normal practice.

Steel Treatment	Max. Stress, tons/sq. in.	Yield Point, tons/sq.in.	Elong. on 2 in., %	Elong. on 8 in., %	Rockwell B
1. Cooled rapidly	25·5	17·6	37·0	25·0	61
Cooled slowly	24·5	15·6	42·5	27·8	56
2. Cooled rapidly	24·4	18·2	36·0	23·5	54·5
Cooled slowly	21·4	16·4	43·5	30·0	51

The variation in mechanical properties is marked and there is a considerable reduction in the ductility of the steel. This is a factor to be considered when comparing continuous single-sheet annealing with the batch system."

Mr. **Pollard**: Some difference would be expected, from the two rates of cooling, in the amount of carbon left in the α solution. With air cooling, slightly more carbon will be kept in solution than with cooling very slowly in a stack, but I should expect such differences to have only a very small effect on the tensile and hardness values, since the total solubility in α iron is of such a low order. On the other hand I should expect a fully annealed sheet to be rather more susceptible to stretcher strains than one with a small amount of carbon in solution in the α iron.

The test results given by Mr. Thomas are of considerable interest, as the differences recorded are quite appreciable, but it is possible that micro-examination might have shown a difference in the resolution of the carbides, as between the air-cooled and pack-cooled samples. If the slowly cooled samples were stacked at 650° C., or rather higher, and the rate of cooling were very slow, a semi-annealing effect might result, which could, at least in part, account for the differences recorded.

Mr. **H. Edwards** (Messrs. John Summers and Sons, Ltd.) presented the paper by himself and Mr. Jones.

Mr. **J. Bromley Davis** (The United Steel Companies, Ltd.) presented his paper.

Mr. **C. L. A. Webb** (Messrs. Stein and Atkinson, Ltd.): It is evident from what the authors of the first paper have said that there are still a number of points in design and operation requiring attention before it can be said that the ideal equipment has been installed. One of the main problems is maintenance, particularly of inner covers and bases for sheet furnaces. The authors, while telling us that mild-steel covers have given an average life of 69 heats, do not say if those covers have been treated with aluminium or some other alloy to improve their life. It is, of course, known that such metallizing treatment improves resistance to oxidation, but it makes straightening and re-welding more difficult. Is it the authors' opinion that inner-cover life has been shortened by the use of the semi-direct-fired furnace as compared with the radiant-tube fired? I recall that a life of over 90 heats was obtained for a number of sheet covers made of chrome-iron sides with mild-steel tops and ends which were used in radiant-tube furnaces. The mild-steel parts failed first.

Would Mr. Edwards care to amplify his remarks on the converted furnaces he mentions, now that he has had longer experience with them?

Our interest in the open-fired cover goes back just over ten years, when the first furnace of this type was installed in this country at a works where mild and high-carbon steels are annealed. These furnaces, which are the multi-stack type, were fitted with sand seals and aluminized mild-steel inner covers. Up to now no furnace lining has been replaced, although some minor

* *Journal of The Iron and Steel Institute*, 1949, vol. 162, May, pp. 79–97.

repairs have been carried out. The lining, which consists of insulating refractory brick, is coated with sillimanite, the burners are tangential, and the flames impinge on the lining. Inner cover life has been very good—generally in line with Mr. Bromley Davis's figures—the covers eventually having to be replaced through mechanical failure. Later designs of this type of furnace have incorporated liquid seals, resulting in better control of annealing and economy in the consumption of atmosphere gas.

The main developments of tin coil and sheet coil annealing have taken place in the U.S.A. These have been in the use of convector plates and high-powered fans, enabling effective edge heating to be obtained and very considerable increases in stacking height. At the same time, open firing instead of radiant-tube firing has been developed, and has received much favourable publicity, but there is, as yet, a comparatively small number installed. In fact, many radiant-tube-fired furnaces are still being built in the U.S.A. on the score of maintenance and high-quality product.

Mention should also be made of the special designs of inner covers, and particularly of the vertical corrugated type which is undoubtedly of value in enabling more rapid heat transfer to be obtained, and thus utilizing more fully the benefits of high stacking, convector plates, and high-powered fans. Inner sleeves are unnecessary.

It should be explained that the horizontal hairpin-shaped radiant tube is operated at a negative pressure, so that if, after long service, any cracks develop, leakage is inwards to the tube, and there is no detrimental effect on the inner cover.

Over a period of ten years, in two of the largest works in the U.S.A., maintenance and replacement costs of radiant tubes were less than 1¼*d.*/ton. Mild-steel inner covers vary from 5*d.* to 9*d.*/ton, depending upon their size, shape, and operation. Experience in this country with radiant tubes, installed in 1939 and fired with coke-oven gas with a higher sulphur content, shows that only now are repairs or small replacements being made.

For the very large furnaces with open firing, an alloy heat-resisting steel inner cover seems to be essential, if a reasonable life is to be obtained, but it is doubtful if maintenance costs will be any less with these than with mild-steel covers and radiant-tube firing. If outputs are pushed up beyond the rated capacity of the equipment, maintenance will be increased.

Mr. Bromley Davis states that the open-fired cover has enabled outputs to be increased by more than 10% and fuel consumption to be reduced by more than 15%. We have had a long experience of both open and radiant-tube firing, and the latest radiant-tube furnaces are heating at least as quickly as direct-fired furnaces.

Although the papers deal with the design of the cover-type furnaces, firing systems, inner covers, and fans, there is little mention of heat recovery. The waste gases leave at temperatures which make heat recovery possible and desirable by the inclusion of a recuperator system to preheat the air for combustion. Economies in fuel consumption of approximately 10% can be effected. Such a recuperator as the "Escher," which is light in weight, can be easily accommodated on the furnace.

For smaller sized units the open-fired cover can be made a complete success, but for the very large units further work is necessary to determine whether the open-fired cover has any advantage over the radiant-tube fired.*

Mr. **J. F. R. Jones** (Messrs. John Summers and Sons, Ltd.): We have not tried the treatment of the surface of the inner cover. We started with mild-steel covers, and they have given a very low cover-cost per ton of steel annealed.

It is probable that the cover life on the two types of furnace will be equally good. The converted in-and-out type of furnace is considered to be quite promising.

I think Mr. Webb is referring to the radiant tube which is exhausting outside the furnace. It is perfectly obvious that in a tube of that description more heat is lost and the thermal efficiency of the furnace cannot be as good as when the gas is exhausting inside the cover.

The question of heat recovery from waste gases has always been in mind, but we have not yet put it into effect. The idea is to use the waste heat in some kind of heat exchanger for space heating of the finishing buildings.

Mr. **Bromley Davis**: There is evidence that the conversion of an existing furnace from radiant-tube to direct firing does give an increased output, if the stock to be heated can absorb the additional heat given by direct firing.

There is a limit to the heat release per unit area that can be obtained from a radiant tube, and there is a limit to the number and size of tubes that can be fitted in a given furnace cover. It appears that a greater total heat release can be obtained from the same furnace cover if direct firing is used in place of radiant-tube firing. When an existing coil-annealing furnace is modified by fitting a high-powered circulating fan together with convector plates, the rate at which heat can be absorbed by the stock is increased and, therefore, the total heat released within the furnace cover should be increased. In these circumstances, a conversion to direct firing would be advantageous. When new furnace equipment is under consideration, it is possible to design a relatively larger furnace cover incorporating radiant tubes having a larger radiating surface, thus giving a greater total heat release.

It is possible to improve the scale-resistant properties of mild steel by treatment of the surface, but as many covers fail, not through scaling, but because of mechanical weakness at the temperature at which they have to work, heat-resisting steel covers might often be found to be an improvement over mild-steel covers.

It may be possible to consider some simple form of recuperator for use with the smaller portable furnace covers, but in the case of the largest furnace covers it is important to keep the cover weight to a minimum so as to avoid difficulties with cranes. For this reason, it may be difficult, if not impossible, to include a recuperator with the larger portable cover furnaces.

Mr. **R. L. Willott**: I think that both authors are to be congratulated on the thought and care which were put into the layout of the shop they describe. Nevertheless, the present bases are most unsatisfactory. I refer particularly to the use of bricks and sand, since the effect of dry sand on loose brickwork soon destroys, in a normal annealing cycle, the original shape of a base. The placing of the inner cover becomes quickly and progressively more difficult, and the formation of a reliable gas seal more hazardous.

Why not use, for the actual furnace base on which the sheets sit, a monolithic structure of, say, a high alumina cement, combined, if necessary, with chrome magnesite, instead of a pile of bricks with a steel plate on top? Such a material might lend itself also to the formation of monolithic sand seals, which would more satisfactorily contain dry sand without losing shape.

We have improved our own technique on one furnace by getting rid of some of the small standard brick previously used. Very much larger shaped blocks have

* Subsequent to this discussion it is learned that a steel combine in the U.S.A., with considerable experience of open-fired furnaces, has placed a large order for annealing equipment of the radiant-tube type.

been built in, and I think Mr. Edwards will agree that has produced a very much better result.

Mr. Webb mentioned liquid seals. If bricks can be dispensed with, and a monolithic surround formed in some way, is there any known liquid which will seal this type of furnace ?

The inner covers which we use at Shotton are made of steel plate $\frac{5}{16}$ in. thick and weigh over 5 tons ; they scale heavily and distort quickly. Because of this distortion, the radiant tubes are damaged and the formation of a gas seal becomes more difficult. Since the sole purpose of an inner cover is to act as a membrane between the outer combustion space and the inner gas atmosphere, surely the weight can be reduced by using very light heat-resisting steel sheet on a suitably designed mild-steel frame. The mild steel could be entirely covered by a heat-resisting skin, and would not be subject to oxidation. If there were any distortion, straightening of such a cover should be comparatively easy—we have found it to be almost impossible with the $\frac{5}{16}$-in. plate at present in use. I believe that some work has already been done in this direction, but there is room for further experiment on these lines.

Mr. **Jones** : The possibility of the liquid seal has not been overlooked. I should be very glad to know from Mr. Webb what he proposes in regard to a liquid seal. I fully agree with Mr. Willott's remarks. Up to the present the mild-steel corrugated inner cover is the best available, and it has worked and given a reasonably cheap annealing cover cost per ton of steel. If we can use a better quality steel of lighter gauge, it will be all to the good, always providing that it does not cost more per ton, including the hidden costs of maintenance.

Mr. **D. Fleming** (The Stourbridge Rolling Mills, Ltd.) : There may be a case for using heat-resisting steel for circular furnace covers, but for rectangular furnaces I agree with Mr. Jones that mild steel is the best, because covers are discarded owing to failure at the welds. Our furnace is smaller than those under discussion, but we have tried plain and corrugated mild and heat-resisting steels from $\frac{1}{16}$ in. to $\frac{1}{4}$ in. thick, and we obtained the best results from $\frac{1}{16}$-in. flat mild steel. These covers can be straightened on the spot with a sledge hammer. There has been reference to covers distorting in a fortnight ; ours start losing shape after two days, but, of course, we are not equipped with radiant tubes.

Mr. **Webb** : We first installed liquid seals on some multi-stack type furnaces in Sheffield for high-carbon steels. The seal is water with an oil film on the top, and it is situated well below the base. The seal is never hot enough to volatilize the oil. We have designed much larger types of coil-annealing furnace with the liquid seals, but we have not yet had an opportunity of testing them. We believe it is possible to design a rectangular sheet base with a liquid seal.

The **Chairman** : I can add to what Mr. Webb has said by saying that a bell-type furnace is operated at Bethlehem of exactly the same shape as the ordinary sheet furnace, with a liquid seal, and the seal is a common seal for the inner cover and the furnace. The inner cover, therefore, has a large skirt on it. We explored the possibility of using that design for multi-stack furnaces, but it would have involved a very complicated series of ' hats ' jointed on a flat bottom.

Mr. **W. F. Gilbertson** (Messrs. Richard Thomas and Baldwins, Ltd.) (Read by the Chairman) : Mr. Bromley Davis refers to the question of the life of the inner covers when using the direct-fired type of furnace. At one plant in the U.S.A., where nothing but direct-fired furnaces are used and where the inner covers are mainly of mild steel, it was found that the inner covers were failing around the top of the cover where the top was welded to the side walls. By pressing the top into a dish form, and thereby making the weld between the top and the side walls of the inner cover about 3 in. down the side wall, much longer life was obtained. This method, of course, means having the top of the cover pressed, which may add somewhat to the original cost of the inner cover.

At one mill in the U.S.A., there was considerable trouble with the convector space plates, which are inserted between the coils, as a result of warping and cracking around the edges. The irregularities were so pronounced that they caused sticker trouble on the edge of the coils in some cases.

At another mill where space plates were used this difficulty was overcome by building all the coils up to a standard outside diameter, which entirely covered the space plate. It was stated that if the space plate was exposed at the edges cracking and warping took place.

Mr. **Bromley Davis** : It does seem that there is room for improvement in the existing design of convector plates. These are normally made of mild steel, but a heat-resisting steel might prove more suitable.

Mr. **H. H. Ascough** (The Steel Company of Wales, Ltd.) : Mr. Bromley Davis mentions the low average value of 1·5 for the thermal conductivity in the coil, taking the steel strip and air gap into consideration. Was the effect of the tension in the coil taken into consideration in his tests ? Obviously the air gap would diminish with increased tension. The gauge variation across the strip would also affect the air gap which, in the case of thick middles, would produce an increasing air gap between the layers and towards the strip edges.

It would be interesting to know what the authors would recommend if they had a free choice in the type of atmosphere gas used, as this has a bearing on the thermal conductivity. Hydrogen or helium have been suggested, but nobody has had the courage to use them.

What are the optimum thicknesses of inner covers that have been found by experiment both in mild steel and alloy steel, and what is the optimum diameter of coil to give the most efficient annealing ? I have read that it is 48 in.

Mr. **Bromley Davis** : The thickness of the strip and the coil tension will have only a small effect on the thermal conductivity for radial heat flow. It has been shown, theoretically, that if the commonly used atmosphere gases were replaced by hydrogen, the thermal conductivity for radiant heat flow could be increased significantly.

Mr. **Edwards** : I should like to refer to Mr. Willott's point about the need for a liquid seal with which it would be possible to reduce the consumption of atmosphere gas by the elimination of the considerable leakage through the sand seal, and the need for improved gas-tight inner covers to reduce further the consumption of atmosphere gas—an expensive item. In this country a wide range of gas atmospheres is used with a measure of success. It is most important to remove sulphur gases to achieve a reasonably white and clean surface finish. It is important to dry, and maybe not quite so important to partially burn, in order to balance the gas composition. In the case of *NX* gas or a nitrogen atmosphere gas, the use of such a composition is very intriguing, because with high percentage nitrogen atmosphere there is a considerable reduction in the toxic effects and the explosion hazard. I have no experience of the use of *NX* gases, but I do not see why excellent results should not be obtained.

We recommend a gas which is relatively low in sulphur —below 30 grains/100 cu. ft.—and with a composition such as we have shown in the paper under column (1) of the data on p. 90.

Mr. **J. Groves** (The Steel Company of Wales, Ltd.): When the temperature of annealing has to be increased, say, 50 to 100° C. above 700° C., the inner-cover question becomes very serious. If something could be found to act as a rib round the base of the big rectangular covers, much of the trouble would be eliminated.

It will be generally agreed that the sticker question is governed by the factors of temperature, roughness of sheet, and height of stack, but as I have to deal with wide cross-rolled cold-reduced sheets and not strip, one big factor has to be added, and that is the shape. The effect of temperature is perhaps over-rated because it is possible to stick at about 650° C. Would the authors modify the figure they give for roughness when considering badly shaped sheets ?

Mr. **Edwards**: The figures for inner-cover life quoted by Mr. Jones and myself relate to the annealing of cold-reduced sheets at temperatures between 650 and 700° C. It has been shown that sheets with a surface roughness below 35 μin. are very sensitive to sticking. I agree wholeheartedly that the shape of the sheets plays an important part in the problem of stickers. When sheets are badly shaped, especially when full in the middle, they are especially prone to sticking.

Mr. **James Lysaght** (The Steel Company of Wales, Ltd.): Would the authors give their opinion and experience, firstly, on automatic control versus hand control on either batch annealing or coil annealing, and, secondly, on the permissible temperature variation over a stack in batch annealing of, say, 100 tons.

Mr. **Edwards**: We would recommend automatic control, if only to ease the tediousness of furnace routine. Unfortunately, automatic control was not possible with dirty gas because of trouble with the valves. With the use of clean gas, we hope to install automatic control throughout the plant. We believe that it will help considerably to control temperature gradients and holding and soaking conditions.

In regard to temperature variation, there are practical difficulties with thermocouples, such as the need for a careful selection of the actual positions; but with couples properly situated, and especially with a base couple well covered, it should be possible to arrive at a temperature variation within ± 10° C., particularly when annealing material which is given relatively long holding and soaking periods.

Mr. **Bromley Davis**: The temperature variation permissible in a coil stack is determined by the metallurgical requirements of the material and the furnace performance. With respect to furnace performance, tests were made on two furnaces, each heating 52-in. dia. tin-plate coils in stacks of four coils. In the furnace using a low-powered circulating fan and no convector plates, a temperature spread of 55° C. was observed after heating for 62 hr. In the other furnace, fitted with a high-powered circulating fan and convector plates, the same temperature spread was observed after heating for only 27 hr.

Mr. **W. E. Hoare** (Tin Research Institute): My reason for raising a point which is not strictly an engineering one, is that the question of rolling oils has been mentioned. Much mild-steel strip is, of course, ultimately finished with either a metallic or non-metallic finish, and the effect of surface characteristics of strip steel on the adhesion and life of applied coatings is important.

Before annealing, mild-steel strip may be electrolytically de-oiled, in which case the choice of the rolling oil is largely governed by oil pick-up characteristics and by power-requirement characteristics. In other cases the rolling oil is left on the strip as it goes to the annealing plant. Other criteria then come into consideration, such as the effect of the residual oil on the final finish of the strip.

The authors indicate that the characteristics of the rolling oil can affect the surface finish, and my own experience tends to confirm this. I would welcome the authors' views on the importance of such effects. We have heard that such factors as temperature and atmosphere can exert major effects on strip surface quality. If oil characteristics also exert a major effect it may be deemed worth while in many cases to de-oil the strip before annealing so that the rolling lubricant can be chosen purely from an 'engineering' standpoint.

Mr. **Edwards**: In the cold reduction of mild steel on modern mills, it is necessary to use a rolling oil of chemical and physical characteristics capable of withstanding high pressures and particularly high temperatures at the nip of the rolls, and of providing the necessary film strength to achieve the reductions planned at the various stands of a mill. A rolling oil is needed which will provide all the necessary properties to achieve the actual cold reduction, but unless there is some means of degreasing and surface treatment before the annealing process, it is necessary to complete the cold reduction stage with a minimum of rolling oil residues on the surface. The oil companies are able to blend and formulate rolling oils with the properties necessary to achieve high reductions, but unfortunately such oils leave excessive residues which contaminate the surface, making it quite unsuitable for many purposes, and aggravate the problem of treatment for enamelling, galvanizing, and tinning. In other words, the best possible blend of rolling oil is required for cold reduction, but at the same time the oil companies are asked to blend in such a manner that the minimum amount of residue is left on the strip so that the steel can be annealed without an expensive surface treatment.

Mr. **H. Kay** (Messrs. Stein and Atkinson, Ltd.): Mr. Bromley Davis does not mention whether he has encountered any oil-fired coil-annealing furnaces. If he has, was there any influence on the maintenance costs of the furnace and the inner covers ?

Although it is better, from the heating point of view, to have no metal interposed between the circulating gases and the edges of the coils, will not the omission of the shrouds cause damage to the edges of the coils ?

Mr. **Bromley Davis**: I see no reason why oil firing should not be used for coil annealing, but in this country, so far as I know, it has been used only on sheet-annealing furnaces.

Concerning the separator plates, I agree with Mr. Kay that, in theory, this type of separator should be better, but I do not know how they will behave in practice.

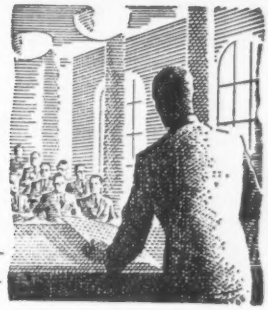
Mr. **R. E. H. Williams** (The Wellman Smith Owen Engineering Corporation, Ltd.): Reference has been made to the life of refractory supporting base blocks. I recollect that some years ago carbon deposition, with subsequent disruption, was a serious problem, and it was felt that blast-furnace inner-wall-quality refractories would be more suitable. Perhaps the modern form of carbon block (or monolith) is worth trying in this position ?

Mr. **C. H. Williams** (Messrs. Stein and Atkinson, Ltd.): Radiant tubes can, of course, be considered as very much the equivalent of electrical elements.

It has been asked whether oil could be used for annealing covers, and although it cannot be applied so easily as gas, it can be used, and is being used successfully both inside radiant tubes and also for open-firing.

Regarding the question of automatic furnace controls, it is generally accepted that these function in a most satisfactory manner on gas-fired furnaces.

DISCUSSION ON PAPERS



DISCUSSION AT MEETINGS · WRITTEN CONTRIBUTIONS

Correspondence on the Paper—

THE MECHANISM OF THE FREEZING OF HORIZONTAL STEEL CASTINGS*

submitted by the Melting and Metallurgical Committee

Mr. **L. K. Royston** (Sheffield) wrote : As I was at one time occupied on one of the group of castings examined by the Melting and Metallurgical Committee I am pleased to see the completed report. During our early training, under the direction of the late Mr. Harry Brearley, we were shown how the various macrostructural characteristics in steel ingots could be reproduced faithfully in stearine-wax models. The fact that stearine wax would reproduce the macro-characteristics of model ingots suggested to me that it might be of use in problems associated with foundry practice.

When engaged on this work of the Committee, we found that by using half-scale moulds of plaster of Paris, and stearine wax, we could reproduce the following characteristics of the steel castings under examination :

(i) Columnar crystal growth from the upper face, which decreased in depth with increasing casting temperature, and equi-axial crystals in the lower half of the section, which became replaced by columnar crystals with further increase in temperature

(ii) The formation of cavities in the castings which, apparently, were not fed efficiently from the feeder head ; the feeder head, being 50% greater in diameter and 75% greater in cross-sectional area than the casting, should have remained molten until the last.

In the wax models the presence of cavities in the casting was accompanied by long columnar crystals in the head, reaching to its central axis, thus forming a network of columnar crystals (dendrites in the steel counterparts) through which the last remaining liquid had difficulty in filtering. This caused a series of cavities in the wax model (cavities and segregation in the steel counterparts) which were more pronounced than would have been anticipated. This type of effect can be observed in a casting with a zone of ' loose ' structure, the restricted feeding being caused by such a columnar crystal growth between the area concerned and the feeding system. Any comments which the Committee may have to make on the macrostructure of the feeder head, with particular reference to the occurrence of prominent columnar crystal growth, will be greatly appreciated.

Another factor, which may affect the prominence of columnar crystal growth, is that the upper face of the mould of a horizontally cast slab will be somewhat cooler, and will give rise to a steeper temperature gradient in the upper half of the casting immediately after the filling of the mould cavity. This would most probably be reflected in the variation of the length of the columnar crystal growth with varying temperature, and its preferential growth in the upper zones.

The various characteristics may be reproduced in a wax mould, which indicates that these properties are fundamental, and may be expected to occur in any system (metallic or otherwise) in which prominent columnar crystal growth occurs. The results of making similar castings in the copper-base alloys would be interesting.

Dr. **W. C. Newell** wrote in reply thanking Mr. Royston for pointing out that the types of structure observed by the Committee in the steel castings studied were not merely of metallurgical application, so that the fundamental mechanism of solidification observed was not dependent upon the metallic properties of the system studied. The Committee had paid only secondary attention to the structure of the feeder heads, but the importance of maintaining free liquid passage between the head and the casting was well appreciated. Recent results had suggested that there might be transport of solid crystals in suspension from the feeder head into the casting, and if so the filtering action of columnar crystals in the head would be enhanced. The points raised would be considered by the Committee when formulating its future programme.

Furthur correspondence on this Report, which will not be presented for verbal discussion, is invited. Contributions should reach the Editor before 20th February, 1950.

* *Journal of The Iron and Steel Institute*, 1949, vol. 162, Aug., pp. 437–450.

N E W S

ANNOUNCEMENTS AND NEWS OF SCIENCE AND INDUSTRY

THE IRON AND STEEL INSTITUTE

Amendments to Bye-Laws

The amendments to the Bye-Laws of the Institute, as adopted at the Annual General Meeting on 27th April, 1949, were allowed by the Privy Council on 27th October, 1949, and are therefore now in force.

The detailed amendments are given in the following Schedule :

(1) Bye-Law 3 *add* thereto the words following :

'It shall also be within the province of the Council to elect Associate Members. The Council shall have power from time to time to make regulations (not inconsistent with anything in these Bye-Laws) for defining the status rights and obligations of Associate Members, for prescribing the ages and qualifications of candidates for Associate Membership and providing for the transfer of Associate Members to ordinary Membership of the Institute including the fixing of an entrance fee to be payable on such transfer and for the waiving of the same at the discretion of the Council.

'Except where the context otherwise requires, the word "Members" where used in these Bye-Laws shall include Members of all classes. Provided that an Associate Member shall not have the right to sign a requisition for a general meeting or to vote at such a meeting or to sign a recommendation for admission to membership or to nominate candidates for election as officers.'

(2) *Delete* Bye-Law 21 and *substitute* therefor the following new Bye-Law :

'21. The entrance fee payable by Members on election shall be such sum, if any, as the Council shall from time to time determine.

'Until otherwise determined by the Council and sanctioned by a resolution passed at either the Annual General Meeting or the General Meeting held in the Autumn of each year, at which notice of intention to propose the same shall be given :

(*a*) The annual subscription payable by ordinary Members shall be £4 4*s*. 0*d*. and by Associate Members shall be £1 15*s*. 0*d*.

(*b*) Any ordinary Member whose subscription is not in arrear may at any time before the 30th June in any year, or if elected after that date in the year, within one month of his election, compound for his subscription for subsequent years on payment according to the following scale :

Up to 29 years of age	£70
From 30 to 39 years of age	£63
,, 40 to 49 ,, ,, ,,	£56
,, 50 to 59 ,, ,, ,,	£49
,, 60 years of age and upwards	£42

Provided that the adoption of this Bye-Law shall not prejudice the validity of any composition duly effected according to the scale set out in the Bye-Law which is hereby revoked.

'Provided also that in the case of any Member or Associate Member who is also a member of the Institute of Metals or of the Institution of Metallurgists or of any other Institute or Society (whether in or out of the United Kingdom) which the Council shall from time to time declare to be a Kindred Society for the purpose of this Bye-Law, the Council shall have power to accept from such Member or Associate Member such reduced subscription under such conditions as it shall from time to time think fit.'

(3) *Delete* from Bye-Law 35 the words "ordinary or honorary" wherever these words occur and also the words "whether ordinary or honorary."

(4) *Delete* from Bye-Law 36 the words "honorary members and honorary vice-presidents" and the words "honorary members or honorary vice-presidents."

Revised Regulations for Associate Members

A class of Associates of The Iron and Steel Institute was established by resolution of the Council on 15th November, 1916. By an amendment to Bye-Law 3 (*see* item (1) in the schedule above) as adopted at the Annual General Meeting, 1949, a class of Associate Members has been established.

The following regulations, to come into force on 1st January, 1950, were adopted by the Council at a meeting held on 9th November, 1949 :

(1) As from 1st January, 1950, all persons who shall previously have been duly elected Associates and who conform to the following regulations shall be Associate Members.

(2) Candidates for election as Associate Members shall be persons not exceeding 25 years of age and they shall have one or other of the following qualifications. They shall be :

(*a*) Students of metallurgy or chemistry or physics or engineering taking courses at a University or University College or a technical college or school.

(*b*) Pupils or apprentices to metallurgists or engineers or in metallurgical or engineering works or laboratories.

(*c*) Persons employed in some practical or scientific capacity in metallurgical or engineering works or laboratories.

(3) No Associate Member shall remain in the class of Associate Member after the 31st day of December next following the day on which he attains the age of 25 except that

(i) In the case of bona-fide full-time students the Council may at their discretion grant exemption from these Regulations, and

(ii) An Associate Member resident in the United States of America or Canada who is a Member of the American Institute of Mining and Metallurgical Engineers, may, under a reciprocal arrangement, remain as Associate Member until the 31st day of December next following the day on which he attains the age of 33, and

(iii) An Associate Member who was an Associate and over the age of 25 on 1st January, 1949, may remain an Associate Member until 31st December next following the day on which he attains the age of 28.

(4) A recommendation for admission as an Associate Member on the form prescribed by the Council shall be forwarded to the Secretary and shall be laid by him before the Council. The recommendation shall be signed by not less than three Members.

(5) Associate Members shall pay an annual subscription of £1 15*s*. 0*d*. per annum except that in the case of any Associate Member who is also a Member of the Institute of Metals or of the Institution of Metallurgists or of any other Institute or Society (whether in or out of the United Kingdom) which the Council shall from time to time declare to be a Kindred Society (*see* Regulation 11) for the purpose of this regulation the annual subscription shall be £1 11*s*. 0*d*. per annum.

(6) Associate Members shall be exempt from payment of an entrance fee.

(7) Associate Members shall have the right to attend all meetings of the Institute and to receive notices thereof and shall have the right to receive publications on the same terms and under the same conditions as ordinary Members.

(8) Associate Members shall not have the right to sign a requisition for a General Meeting or to vote at such a Meeting or to sign a recommendation for admission to membership or to nominate candidates for election as officers.

(9) An Associate Member on attaining the above age limits, or at any previous date if over 21 years of age, may on application be transferred by the Council to ordinary membership, if he is qualified to be an ordinary Member under Bye-Law 3.

(10) Associate Members on transfer to the class of ordinary Members shall pay thereafter the usual annual subscription of ordinary Members but shall be exempt from payment of any entrance or transfer fee.

(11) The Council shall from time to time by resolution decide what Institutes or Societies shall be declared to be Kindred Societies for the purpose of these regulations; the following Institutes and Societies are hereby declared to be such Kindred Societies:

Foreign Societies

American Institute of Mining and Metallurgical Engineers
American Iron and Steel Institute
American Society for Metals
Association for Czechoslovak Mining and Metallurgical Engineers (Czechoslovakia)
Association des Ingénieurs sortis de l'Ecole de Liège (Belgium)
Associazione Italiana di Metallurgia (Italy)
Association Luxembourgeoise des Ingénieurs et Industriels (Luxemburg)
Australian Institute of Metals
Canadian Institute of Mining and Metallurgy
Indian Institute of Metals
Instituto del Hierro y del Acero (Spain)
Jernkontoret (Sweden)
Koninklyk Instituut van Ingenieurs (Holland)
Norsk Metallurgisk Selskap (Norway)
Société Française de Métallurgie (France)
Société Royale Belge des Ingénieurs et des Industriels (Belgium)
Verein Schweizerischer Maschinen-Industrieller (Switzerland)

Affiliated Local Societies

Cleveland Institution of Engineers
Ebbw Vale Metallurgical Society
Leeds Metallurgical Society
The Lincolnshire Iron and Steel Institute
The Manchester Metallurgical Society
The Newport and District Metallurgical Society
Sheffield Society of Engineers and Metallurgists
Staffordshire Iron and Steel Institute
The Swansea and District Metallurgical Society

Annual General Meeting, 1950

The Annual General Meeting of The Iron and Steel Institute will be held in London on Wednesday and Thursday, 26th and 27th April, 1950.

Sidney Gilchrist Thomas Centenary

Mr. JAMES MITCHELL, C.B.E., Honorary Treasurer, will give a Special Commemorative Lecture to mark the centenary of the birth of Sidney Gilchrist Thomas.

The Lecture will be delivered in London on the evening of Wednesday, 26th April, 1950, the first day of the Annual General Meeting of the Institute.

Further particulars will be announced later.

Foundry Prize

Mr. PHILIP SAMUEL, who recently obtained the degree of B.Met. (Founding), with first-class honours, at Sheffield University, has been awarded the Foundry Prize of The Iron and Steel Institute.

Blast-Furnace Practice

A member wishes to purchase a good second-hand copy of "Blast Furnace Practice" (3 vols.) by F. Clements (E. Bean and Co.). Would any reader willing to dispose of a copy please send particulars to the Editor at 4 Grosvenor Gardens, London, S.W.1.

NEWS OF MEMBERS

▶ Mr. W. A. ADAM has left the Anglo-Argentine Iron Co., Ltd., of Buenos Aires, and has joined Messrs. John Summers and Sons, Ltd., Shotton, Chester.

▶ Mr. J. B. R. BROOKE is with Northern Construction (Rhodesia), Ltd., in Salisbury, Southern Rhodesia.

▶ Mr. WAYNE L. COCKRELL is leaving New Jersey, U.S.A., to take up duties with the Economic Co-operation Administration Mission to Korea, at San Francisco, California.

▶ Dr. B. D. CULLITY has left the U.S. Office of Naval Research in London, where he was a scientific liaison officer, to take up an appointment as Assistant Professor of Metallurgy at the University of Notre Dame, Notre Dame, Indiana, U.S.A.

▶ Dr. C. H. DESCH, F.R.S., was installed as President of the Newcomen Society, on 9th November, 1949.

▶ Mr. GEORGE ELLIOTT has been appointed to the Board of Directors of Messrs. James Dougall and Sons, Ltd., Bonnybridge, Scotland.

▶ Mr. G. L. EVANS has left the Municipal Technical College, Swansea, and is now in the Metallurgy Department, Royal School of Mines, South Kensington, London.

▶ Monsieur E. S. L. FOULD, Directeur Général de la S.A. des Hauts Fourneaux, Forges et Aciéries de Pompey,

has been made a Chevalier de la Légion d'Honneur.

▶ Mr. H. S. GUPTA has been appointed Assistant Professor, Department of Metallurgy, at the Government Engineering College of Central Provinces and Berar, Jubbulpore.

▶ Mr. B. K. MAHENDRA has changed his name to B. K. Sood. He has been awarded the degree of B.Sc.(Met.) of the Benares Hindu University.

▶ Mr. JAMES M. MOWAT, Chief of Research at the Parkhead Steel Works of Messrs. William Beardmore and Co., Ltd., has been appointed a Special Director.

▶ Mr. V. A. PHILLIPS has returned from the Hammond Metallurgical Laboratory, Yale University, U.S.A., and has joined the Cavendish Laboratory, Cambridge, working under Dr. E. Orowan.

▶ Dr. G. V. RAYNOR has been appointed Professor in Metal Physics at the University of Birmingham.

▶ Mr. H. A. SNOW has taken up the appointment of Assistant to the Technical Director of Messrs. Edgar Vaughan and Co., Ltd., Birmingham.

▶ Mr. J. B. WAGSTAFF has left Hydrocarbon Research Inc., New Jersey, U.S.A., to take up the position of Combustion Engineer at the Central Research Laboratory of the United States Steel Corporation, Delaware, U.S.A.

▶ Mr. D. R. F. WEST has been awarded the degree of Ph.D. of London University, and the Diploma of the Imperial College of Science and Technology. He is now a scientific officer at the Armament Research Establishment, Woolwich.

▶ Mr. H. WILLIAMS has left The Steel Company of Wales, Ltd., to become Assistant Plant Manager, Powdered Metals Division, of Murex Ltd., Rainham, Essex.

Obituary

Miss D. P. CLAPHAM, of the Ministry of Supply, Fort Halstead, Sevenoaks, on 29th June, 1949.

Mr. P. A. E. Armstrong, Vice-President of the Allegheny Ludlam Steel Corporation, Pittsburgh, Pa., on 7th August, 1949.

CONTRIBUTORS TO THE JOURNAL

L. D. Jaffe, Sc.D., M.A.I.M.E.—Head of the Physical Metallurgy Branch, Watertown Arsenal Laboratory, Watertown, Mass., U.S.A. Dr. Jaffe received degrees from the Massachusetts Institute of Technology and from Harvard University. In 1940 he joined the staff of Watertown Arsenal Laboratory as a junior metallurgist, and has been associated with that Laboratory ever since.

F. László, Dr. Ing.—Dr László is on the teaching staff of the Engineering School of the University of Melbourne, Australia. He was born in Sopron, Hungary, in 1894. He graduated in mechanical and electrical engineering at the Budapest Engineering School, and obtained the degree of Dr.Ing. at the Darmstadt Engineering School, Germany, both with first-class honours. Until 1938, Dr. László was in industry, firstly, for one year in Hungary, and then for twelve years as head of the Metallurgical Laboratory of the Turbine Factory of Messrs. Siemens-Schuckertwerke A.G., in Germany. For a further five years he was head of the Central Metallurgical Laboratory of that company. He took up his present position in 1939.

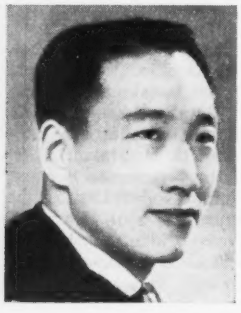
Tsun Ko, Ph.D.—Lecturer in the Metallurgy Department of the University of Birmingham. Dr. Ko was educated at the Hopei Institute of Technology, Tientsin, and at the National Wuhan University, Wuchang, China, where he obtained a first-class honours degree in Chemistry, in 1938. After serving with the Chinese Ministry of Economic Affairs as a scientific officer, and having obtained practical experience at the Tata Iron and Steel Works, Jamshedpur, India, he received an I.C.I. Research Fellowship, and entered the University of Birmingham in 1945. He was awarded the degree of Ph.D. in 1948 for researches into the phenomenon of recrystallization and on the effect of high temperature heating on steel. He took up his present position in October, 1948.

A. Hartley, B.Sc.—A member of the Steel Castings Division of the British Iron and Steel Research Association. Mr. Hartley was educated at the Sowerby Bridge Grammar School and at Leeds University where he was awarded an honours degree in metallurgy in 1942. He subsequently carried out research work for the Alloy Steels Research Committee, on the investigation of the oxidation and also the overheating of steels. He took up his present position in 1947.

J. Nutting, B.Sc., Ph.D.—Demonstrator in Metallurgy in the University of Cambridge. Dr. Nutting was educated at Mirfield Grammar School, Yorkshire, and at the University of Leeds. He obtained the degree of B.Sc. with first-class honours in metallurgy in 1945, and the degree of Ph.D., in 1948, for a thesis on the overheating and burning of steel. From 1948–49 he was employed by the British Iron and Steel Research Association at the Cavendish Laboratory, Cambridge, working on the electron microscopy of metals. He took up his present position in October, 1949.

A. Preece, B.Sc., M.Sc., F.I.M.—Professor of Metallurgy at King's College, University of Durham. Mr. Preece was educated at Port Talbot County School and at Swansea University. He has been Assistant Metallurgist at the Pressed Steel Company of Great Britain, Research

F. László



T. Ko

L. D. Jaffe

J. Nutting

Assistant to Principal C. A. Edwards, F.R.S., at Swansea University, and a Research Officer at the South Wales Siemens Steel Research Association. Before his present appointment in 1948, Mr. Preece was Lecturer and Reader in Metallurgy at Leeds University. He is a Past-President of the Leeds Metallurgical Society and a member of many technical committees.

R. M. J. Withers, M.A., A.Inst.P.—Scientific Officer in the Instruments Section of the Physics Department of the British Iron and Steel Research Association. Mr. Withers was educated at Peter Symonds School, Winchester, and at Christ's College, Cambridge, where he graduated with honours, taking the Natural Sciences Tripos, Part I. In August 1944 he joined the staff of the Royal Observatory, Greenwich, where he was engaged in research work in the Time Department, specializing in the development of electronic and optical instruments. He took up his present position in September 1947.

IRON AND STEEL ENGINEERS GROUP

The TWELFTH MEETING of the Engineers Group will be held at 4 Grosvenor Gardens on Thursday, 23rd March, 1950. Details of the papers to be presented and discussed will be given in later issues of the *Journal*.

AFFILIATED LOCAL SOCIETIES

Sheffield Society of Engineers and Metallurgists

Professor H. W. Swift, D.Sc. has been elected President of the Society for the Session 1950–51.

INSTITUTION OF METALLURGISTS

Examination Results

The Examinations of the Institution of Metallurgists were held from 29th August to 6th September, 1949, in Birmingham, Glasgow, London, and Swansea. The following candidates were successful :

Licentiateship—J. E. S. Cockle (Beckenham) ; E. S. Harwood (Birmingham) ; J. P. Moore (London) ; T. R. Savage (Sunderland) ; E. Ward (Doncaster) ; B. R. Watson-Adams (Ashford).

Associateship (From Licentiateship)—F. R. H. Allon (Southampton) ; H. B. Dunthorne (Slough) ; W. G. Hull (Slough) ; G. R. Morton (Kettering).

Fellowship (From Associateship)—F. C. Evans (Gerrards Cross).

The following candidates satisfied the Examiners in the Licentiateship Examination and Associateship Examination, respectively, but their election is deferred until they have been able to comply with the regulations concerning age and experience :

A. P. Miodownik (London) Candidate for the Licentiateship Examination ; A. Wilson (Glasgow) (Licentiate) Candidate for the Associateship Examination.

EDUCATION

Lectures at Birmingham University

At the request of members of the Institute of Physics and the Institute of Metals the University of Birmingham Extra Mural Department, in co-operation with the Departments of Metallurgy and Chemistry, is arranging a course of lectures for research workers in industry. This series of twenty-four lectures, the most comprehensive which the University has so far provided extra-murally for industry, commences on 14th January, 1950, at the University, Edmund Street. The course has the general title 'Modern Views of the Solid State.' An introductory section on the general theoretical background of wave mechanics (F. R. N. Nabarro) leads on to more specialized courses dealing with non-metallic solids (J. E. B. Randles), and with metals and alloys (B. A. Bilby).

Enquiries about the course should be sent to the Director of the Extra Mural Department, Edmund Street, Birmingham 3.

Mond Nickel Fellowships

The Mond Nickel Fellowships Committee has pleasure in announcing the following awards for 1949 :

Mr. J. MONAGHAN (Stewarts and Lloyds, Ltd.) : To study the method of control and administration of basic open-hearth operation and practice in the steel industry in the United States of America.

Mr. R. STEWARTSON (The United Steel Companies Ltd.) : To study the design and operation of modern hot-rolling-mill plant in the United States of America.

Full particulars of Fellowships, which are awarded annually by The Mond Nickel Co., Ltd., can be obtained from the Secretary, Mond Nickel Fellowships Committee, 4 Grosvenor Gardens, London, S.W.1.

Courses for Workpeople's Representatives

Full-time residential courses in industrial administration for workpeople's representatives are now being run at the Birmingham Central Technical College with the approval of the Ministry of Education.

The scheme was proposed by the Birmingham Local Education Authority in co-operation with the T.U.C. and Trade Unions in the area, and will provide for the special study of factory organization and management, industrial relations and industrial law, motion study and time study, and the use of English both written and spoken.

Students will be sponsored by Trade Unions or other similar organizations, and the Municipal and General Workers Union have already nominated sufficient officials to fill the first course, as well as the second course which has just opened. Three other courses will be held early in the New Year.

NEWS OF SCIENCE AND INDUSTRY

Professor F. L. Meyenberg

Professor Meyenberg, well-known production engineer, died on 1st October, 1949, whilst in Frankfurt. He was born in 1875 and received his education in Hanover and Berlin. After graduating with honours as Dipl.Ing. he became assistant to the Director of the Institute of Technical Physics at Göttingen University.

In 1900 Professor Meyenberg left the academic field, and for the next twenty years or so he held various industrial positions in Germany. He returned to University life in 1923, becoming Professor of Industrial Administration at Brunswick University.

With the advent of the Nazi regime he came to England in 1933, eventually becoming a naturalized British subject. After a period on the technical staff of the British Iron and Steel Federation, Professor Meyenberg joined the research staff of The United Steel Companies, Ltd., acting as an efficiency engineer with the Appleby Frodingham Steel Co., and later with Messrs. Samuel Fox and Company.

He came to London in 1941, and from that time until his death he concerned himself with writing technical

books, including *The Economic Control of Iron and Steel Works*, and with editorial work on the staff of the *Engineers' Digest*.

Professor Meyenberg was a member of The Institution of Mechanical Engineers and an honorary member of the Institute of Economic Engineering.

Mr. J. P. D. Coleman

Mr. Coleman, a Director of Messrs. Wild-Barfield Electric Furnaces, Ltd., died on 20th November, 1949. He joined the National Telephone Company in 1896, under the late Mr. E. P. Barfield, and after a period at sea he entered the cable department of Messrs. Siemens Brothers and Co., in 1910.

During the First World War Mr. Coleman served with the Royal Engineers. In 1919 he rejoined Mr. Barfield and became Works Manager of the new company formed with Mr. L. W. Wild. He was appointed to the Board as Works Director in 1933, and five years later was made a Director of G.W.B. Electric Furnaces, Ltd. He retired from these positions in 1948.

Corrigenda

The Corrosion Test in Relation to Cold Rolling

The following corrections and additions should be made to the paper by N. H. Polakowski (*Journal of The Iron and Steel Institute*, 1949, vol. 163, Nov., pp. 250–276):

Page 250, left-hand column, second paragraph, last line, should read: "$+10$ and -35%."

Pages 251 and 252, Tables I and II, the values are in tons/sq. in.

Page 254, left-hand column, lines 3, 4, 7, and 8, the curve numbers should be 1′, 1, 1, and 1′, respectively.

Page 259, Table III, the analysis of steel 018-a is now given as: C 0·18%, Mn 0·64%, Si 0·04%, S 0·026%, P 0·009%.

Page 260, Fig. 18, second and third lines of caption should read: ". . . values). Cu50-a copper, . . . "

Page 265, Fig. 24, the equation should read:

$$p = \frac{P}{\sqrt{Rb(b\delta + 0{\cdot}000334P)}}$$

Page 266, Fig. 25, the reference 15 should be added after the name of Ford in the caption.

Page 274, Fig. 35, curve *d*, the number of pass should be 4.

DIARY

8th Jan.—LEEDS METALLURGICAL SOCIETY—Film on "*The Bragg Soap Bubble*"—The Department of Chemistry, University of Leeds, 7.0 P.M.

11th Jan.—NORTH WALES METALLURGICAL SOCIETY (Joint Meeting with the Liverpool Metallurgical Society)—"*Recent Developments in the Study of the Flow of Steel*," by M. W. Thring—County Primary School, Plymouth Street, Shotton, Chester, 7.15 P.M.

14th Jan.—SWANSEA AND DISTRICT METALLURGICAL SOCIETY—(Joint Meeting with The Iron and Steel Institute)—"*The Conversion to Oil-Firing at the Park Gate Works*," by D. F. Marshall and H. C. White—Central Library, Alexandra Road, Swansea, 6.30 P.M.

16th Jan.—SHEFFIELD SOCIETY OF ENGINEERS AND METALLURGISTS — Presidential Address — Royal Victoria Station Hotel, Sheffield, 6 P.M.

18th Jan.—MANCHESTER METALLURGICAL SOCIETY—"*Photo-Elasticity*," by J. D. Ward—The Engineers' Club, Albert Square, Manchester, 6.30 P.M.

19th Jan.—CLEVELAND INSTITUTION OF ENGINEERS (Joint Meeting with The Iron and Steel Institute)—Talk on steelworks in the U.S.A., by Sir Charles Goodeve, F.R.S.—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 6.30 P.M.

20th Jan.—WEST OF SCOTLAND IRON AND STEEL INSTITUTE—"*Developments in Low Alloy Structural Steels*," by I. M. Mackenzie and J. M. Pow—39 Elmbank Crescent, Glasgow, 6.45 P.M.

25th Jan.—THE INSTITUTE OF WELDING—"*The Welding of Light Alloys in Structural and Similar Applications*," by W. K. B. Marshall—The Institution of Civil Engineers, Great George Street, London, S.W.1, 6.0 P.M.

27th Jan.—SHEFFIELD METALLURGICAL ASSOCIATION (Refractories Group)—"*Carbon as a Refractory Material for the Iron and Steel Industry*," by K. W. Cowling—198 West Street, Sheffield, 9, 7.0 P.M.

28th Jan.—SWANSEA AND DISTRICT METALLURGICAL SOCIETY—Annual Dinner—Mackworth Hotel, Swansea, 6.30 P.M.

TRANSLATION SERVICE

(The previous announcement was made in the December, 1949, issue of the *Journal*, p. 458).

TRANSLATIONS AVAILABLE

No. 391 (German). P. REICHARDT: "The Influence of the Throat Diameter on the Behaviour of Fines in the Blast-Furnace." (*Stahl und Eisen*, 1949, vol. 69, July 21, pp. 503–508). (Translated by Mr. R. Sewell and made available through the courtesy of the Research and Development Department, The United Steel Companies, Ltd., Stocksbridge, near Sheffield).

No. 392 (German). H. HERBST: "The Alternating Bend Test for Haulage Cable Wires." (*Glückauf*, 1924, vol. 60, Nov. 29, pp. 1111–1120). (Translation made available through the courtesy of the British Iron and Steel Research Association. Members of that Association may obtain copies free of charge on application to B.I.S.R.A.).

TRANSLATIONS IN COURSE OF PREPARATION

(Russian). I. M. CHIZHIKAV: "Evaluation of the Plasticity of Metals and Alloys as Regards Conditions of Hot Shaping by Pressure." (*Zavodskaya Laboratoriya*, 1949, vol. 15, Feb., pp. 191–199).

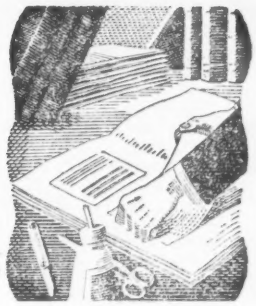
(German). W. LUEG: "The Determination of Spread in the Hot-Rolling of Steel with Smooth Rolls." (*Archiv für das Eisenhüttenwesen*, 1949, vol. 20, Jan.–Feb., pp. 59–68). (Translation made available through the courtesy of the British Iron and Steel Research Association. Members of that Association may obtain copies free of charge on application to B.I.S.R.A.).

(German). E. A. SPENLÉ: "The Cogging Mill in the Special-Steel Rolling Mill." (*Stahl und Eisen*, 1949, vol. 69, June 23, pp. 443–449; discussion, pp. 449–450).

CHARGES FOR COPIES OF TRANSLATIONS—The charge for the above translations is £1 for the first copy and 10*s*. for each additional copy of the same translation. Requests should be accompanied by a remittance. These translations are not available on loan from the Joint Library.

TRANSLATIONS PREPARED AT MEMBERS' REQUEST—Members requiring translations of foreign papers are invited to communicate with the Secretary, who will ascertain whether they can be prepared for inclusion in the Series.

ABSTRACTS OF CURRENT LITERATURE *and* BOOK NOTICES



IRON AND STEEL MANUFACTURE AND RELATED SUBJECTS

CONTENTS

	PAGE
List of Periodicals	97
Ores—Mining and Treatment	102
Fuel—Preparation, Properties, and Uses	103
Temperature Measurement and Control	104
Refractory Materials	104
Blast-Furnace Practice and Production of Pig Iron	104
Direct Processes	105
Production of Steel	105
Foundry Practice	106
Heating Furnaces and Soaking Pits	107
Heat-Treatment and Heat-Treatment Furnaces	107
Forging, Stamping, Drawing, and Pressing	108
Rolling-Mill Practice	109
Lubricants and Lubrication	109
Welding and Flame-Cutting	110
Machining and Machinability	110
Cleaning and Pickling	111
Protective Coatings	111
Powder Metallurgy	113
Properties and Tests	114
Metallography	122
Corrosion	123
Analysis	125
Book Notices	125
New Publications	128

LIST OF PERIODICALS EXAMINED BY THE ABSTRACTING SERVICE

ASEA : s Tidning (Västerås)
Academia Brasileira de Ciencias (Rio de Janeiro)
Anais
Académie Bulgare des Sciences (Sofia)
Comptes Rendus
Académie Polonaise des Sciences Techniques (Warsaw)
Annales
Académie des Sciences (Paris)
Comptes Rendus
Accademia Nazionale dei Lincei (Rome)
Rendiconte (Fisiche Matematische e Naturali)
Acero y Energia (Barcelona)
Acid Open-Hearth Research Association (Pittsburgh)
Bulletin
Acier (Paris)
Acta Crystallographica (London)
Acta Mineralogica Petrographica (Szeged, Hungary)
Acta Polytechnica (Stockholm)
Advancement of Science (London)
Aero Research Technical Notes (Duxford, Cambs)
Aeroplane (London)
Affärsekonomi (Stockholm)
Aircraft Engineering (London)
Aircraft Production (London)
Alloy Casting Bulletin (New York)
Alloy Metals Review (Widnes, Lancs.)
Alluminio (Milan)
Aluminium (Budapest)
Aluminium Courier (London)
Aluminium Development Association (London)
Information Bulletin
Aluminium Laboratories, Ltd. (Montreal)
Abstract Bulletin
American Ceramic Society (Columbus, Ohio)
Bulletin
Journal
American Chemical Society (Washington, D.C.)
Journal
American Electroplaters' Society (Jenkintown, Pa.)
Proceedings
Research Reports
American Foundryman (Chicago)
American Foundrymen's Society (Chicago)
Transactions
American Gas Association (New York)
Monthly
Natural Gas Department, Proceedings
Proceedings
American Gas Journal (New York)
American Institute of Electrical Engineers (New York)
Transactions
American Institute of Mining and Metallurgical Engineers (New York)
Blast-Furnace and Raw Materials Committee, Proceedings
Contributions
Electric Furnace Steel Conference, Proceedings
National Open-Hearth Committee, Proceedings
Technical Publications
Transactions
American Iron and Steel Institute (New York)
Annual Statistical Reports
Contributions to the Metallurgy of Steel
National Emergency Steel Specifications.
Yearbook
American Journal of Physics (New York)
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American Society of Civil Engineers (New York)
Proceedings
Transactions
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Transactions
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Transactions
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Journal
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Bulletins
Proceedings
Standards
Symposia
American Water Works Association (New York)
Journal
American Welding Society (New York)
Miscellaneous Publications
American Zinc Institute (New York)
Journal
Anales de Fisica y Química (Madrid)
Anales de Ingenieria (Bogota, Colombia)
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Analyst (London)
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Analytical Chemistry (Washington, D.C.)
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Mémoires
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Reports
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LIST OF PERIODICALS

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Archiv für Metallkunde (Heidelberg and Berlin)
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Asociación Química Argentina (Buenos Aires)
Anales
Associação Brasileira de Metais (São Paulo)
Boletim
Associação Química do Brasil (Rio de Janeiro
Anais
Boletim
Association des Industriels de Belgique (Brussels)
Prévention des Accidents
Association des Ingénieurs de la Faculté Polytechnique de Mons (Mons)
Publications
Association of Iron and Steel Engineers (Pittsburgh)
Yearly Proceedings
Association Technique de l'Acier et des Métaux Non-Ferreux (Paris)
Bulletin d'Information et de Documentation
Australasian Engineer (Sydney)
Australasian Institute of Mining and Metallurgy (Melbourne)
Proceedings
Australia—Defence Research Laboratories (Maribyrnong)
Circulars
Plating notes
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BEAMA Journal (London)
B.H.P. Review (Melbourne)
Ball Bearing Journal (Luton)
Báňsky Obzor (Prague)
Bányászati és Kohászati Lapok (Budapest)
Battelle Memorial Institute (Columbus, Ohio)
Miscellaneous Publications
Bell Laboratories Record (New York)
Bell System Technical Journal (New York)
Bell Telephone Magazine (New York)
Bell Telephone System (New York)
Technical Publications
Berg- und Hüttenmännische Monatshefte (Vienna)
Betrieb und Fertigung (Vienna)
Birmingham Metallurgical Society (Birmingham)
Journal
Blad för Bergshandteringens Vänner (Stockholm)
Blast-Furnace and Steel Plant (Pittsburgh)
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German Division (Document Unit)
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Brass Forging Association (New York)
Non-Ferrous Forging Digest
British Abstracts (London)
British Cast Iron Research Association (Birmingham)
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British Ceramic Abstracts (Stoke-on-Trent)
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Transactions
British Coal Utilisation Research Association (London)
Monthly Bulletin
Quarterly Gazette
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Technical Bulletin
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British Intelligence Objectives Sub-Committee (London)
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Technical Reports
British Non-Ferrous Metals Research Association (London)
Bulletin
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Bulletin
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Journal
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British Standards
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Bulletin
Log
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Reports of Investigations
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Transactions
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Canning Trade (Baltimore, Md.)
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Bulletin Analytique
Circulaires d'Informations Techniques
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Bulletin
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Chemical Age (London)
Chemical Engineering (New York)
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Proceedings
Chemical Engineering Mining Review (Melbourne)
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Coke and Gas (London)
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Division of Aeronautics, Reports
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Journal

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Bulletin
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Journal
Freyn Design (Chicago)
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Communications
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Quarterly Journal
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Bulletin
Memoirs
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Memoirs
Records
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Bulletins
Circulars
Professional Papers
Geologiska Föreningens (Stockholm)
Förhandlingar
Gjuteriet (Stockholm)
Glównego Instytutu Metalurgii i Odlewnictwa (Gliwice)
Prace Badawcze
Glückauf (Essen)
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Congrès

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Miscellaneous Publications
Hutnické Listy (Prague)
Hutnik (Warsaw)

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Imperial Smelting Corporation (Avonmouth)
Monthly Abstracts
Index Aeronauticus (London)
Indian Academy of Sciences (Bangalore)
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Journal
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Indian Institute of Metals (New Delhi)
Indian Institute of Science (Bangalore)
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Indiana Academy of Science (Greencastle, Ind.)
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Industria y Química (Buenos Aires)
Industrial Chemist (London)
Industrial Diamond Information Bureau (London)
Bibliographies
Industrial Diamond Review (London)
Industrial and Engineering Chemistry (Washington, D.C.)
Industrial Finishing (Leighton Buzzard, England)
Industrial Heating (Pittsburgh)
Industrial Heating Engineer (London)
Industrial and Mining Standard (Melbourne)
Industrie und Technik (Vienna)
Industritidningen Norden (Stockholm)
Ingenieria e Industria (Buenos Aires)
Ingénieur Chimiste (Brussels)
Ingenioren (Copenhagen)
Ingeniors Vetenskaps Akadamien (Stockholm)
Handlingar
I.V.A.
Ingeniörsklubbens i Falun Förhandlingar (Falun, Sweden)
Institut National de Recherches Technologiques de Roumanie (Bucharest
Bulletin
Institut de Recherches de la Sidérurgie (Paris)
Publications
Institute of Australian Foundrymen (Melbourne)
Proceedings
Institute of British Foundrymen (Manchester)
Journal
Proceedings
Institute of Fuel (London)
Journal
Institute of Marine Engineers (London)
Transactions
Institute of Metals (London)
Journal
Institute of Petroleum (London)
I.P. Review
Journal
Institute of Vitreous Enamellers (London)
Annual Proceedings
Institute of Welding (London)
Transactions
Institution of Automotive and Aeronautical Engineers (Melbourne)
Journal

LIST OF PERIODICALS

Institution of Chemical Engineers (London)
Quarterly Bulletin
Transactions
Institution of Civil Engineers (London)
Journal
Institution of Electrical Engineers (London)
Journal.
Institution of Engineers, Australia (Sydney)
Journal
Institution of Engineers and Shipbuilders in Scotland (Glasgow)
Transactions
Institution of Gas Engineers (London)
Papers
Transactions
Institution of Heating and Ventilating Engineers (London)
Journal
Institution of Locomotive Engineers (London)
Journal
Institution of Mechanical Engineers (London)
Journal
Proceedings
Proceedings of the Automobile Division
Institution of Metallurgists (London)
Bulletin
Institution of Mining Engineers (London)
Transactions
Institution of Mining and Metallurgy (London)
Bulletin
Transactions
Institution of Naval Architects (London)
Transactions
Institution of Production Engineers (London)
Journal
Instituto Geologico y Minero de España (Madrid)
Boletin
Mapa Geologica
Memorias
Notas y Comunicaciones
Instituto del Heirro y del Acero (Madrid)
Instituto Nacional de Tecnologia (Rio de Janeiro)
Reports
Instituto de la Soldadura (Madrid)
Boletin de Información
Publicaciones
Instrument Practice (London)
International Acetylene Association (New York)
Oxy-Acetylene Committee Publications
Proceedings
International Institute of Welding (Paris)
Bibliographical Bulletin of Welding
Iron Age (New York)
Iron and Coal Trades Review (London)
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Iron and Steel Engineer (Pittsburgh)
Ironmonger (London)
Istanbul University (Istanbul)
Revue de la Faculté des Sciences

Jassy—Institut Polytechnique de Jassy (Jassy, Rumania)
Jernindustri (Oslo)
Jernkontorets Annaler (Stockholm)
Johnson Matthey and Co. Ltd. (London)
Bulletin
Joint Intelligence Objectives Agency (London)
Reports
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Journal of Dental Research (Baltimore, Md.)
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Journal du Four Électrique (Paris)
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Journal of Petroleum Technology (Dallas, Texas)
Journal de Physique et le Radium (Paris)
Journal of Scientific Instruments (London)
Junior Institution of Engineers (London)
Journal

Kgl. ung. Palatin-Joseph-Universität für technische und Wirtschaftswissenschaften Fakultät für Berg-, Hütten-, und Forstwesen, Sopron (Sopron, Hungary)
Mitteilungen der Berg- und Hüttenmännischen Abteilung
Kullagertidningen (Gothenburg, Sweden)
Kyushu University, Faculty of Engineering (Kyushu, Japan)
Memoirs

Lastechniek (The Hague)
Leeds Philosophical and Literary Society (Leeds)
Proceedings
Leningrad M. I. Kalinin Polytechnic Inst. (Leningrad)
Transactions
Light Metal Age (Chicago)
Light Metals (London)
Lincolnshire Iron and Steel Institute (Scunthorpe)
Proceedings
Linde Tips (New York)
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Industrial Bulletin
Liverpool Engineering Society (Liverpool)
Transactions
Lukens Steel Company (Coatesville, Pa.)
Clad News

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Machinery (Brighton, Sussex)
Machinist (London)
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Malleable Iron Facts (Cleveland, Ohio)
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Manchester Association of Engineers (Manchester)
Transactions
Maquinas e Metais (Lisbon)
Massachusetts Institute of Technology (Cambridge, Mass.)
Proceedings of Summer Conference on Spectroscopy
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Mechanical Engineering (New York)
Mechanical Handling (London)
Mechanical World (London)
Melbourne University (Melbourne)
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Mémorial de l'Artillerie Française (Paris)
Mémoires
Metal Bulletin (London)
Metal Finishing (New York)
Metal Finishing Association (London)
Publications
Metal Industry (London)
Metal Powder Association (New York)
Proceedings
Metal Powder News (New York)
Metal Processing Bulletin (London)
Metal Progress (Cleveland, Ohio)
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Metalen (The Hague)
Metalloberfläche (Munich)
Metallurgia (Manchester)
Metallurgia Italiana (Milan)
Métallurgie (Paris)
Metallwirtschaft (Berlin)
Metals (New York)
Metals Review (Cleveland)
Metalurgia y Electricidad (Madrid)
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Metropolitan-Vickers Gazette (Manchester)
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Microtecnic (Lausanne)
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Transactions
Mining Institute of Scotland (Glasgow)
Transactions
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Minnesota University, Engineering Experiment Station (Minneapolis, Minn.)
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Missouri University School of Mines and Metallurgy (Rolla, Mo.)
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Modern Metals (Chicago)
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Bulletin Series
National Smoke Abatement Society (London)
Smokeless Air
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Neue Giesserei (Düsseldorf)
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Transactions
Nickel Bulletin (London)
Nickel Cast Iron News (New York)
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North-East Coast Institution of Engineers and Shipbuilders (Newcastle-on-Tyne)
Transactions
Nova Scotian Institute of Science (Halifax, Nova Scotia)
Proceedings
Nucleonics (New York)

Office Technique pour l'Utilisation de l'Acier (Paris)
Acier
Ohio State University, Engineering Experiment Station (Columbus, Ohio)
Bulletin
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Reports
Optical Society of America (New York)
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Ossature Métallique (Brussels)
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Anzeiger
Österreichischer Maschinenmarkt und Elektrowirtschaft (Vienna)
Overseas Engineer (London)

Peintures, Pigments, Vernis (Paris)
Petroleum Technology (London)
Philippine Geologist (Manila)
Philips' (N.V.) Gloeilampenfabrieken, Laboratoria (Eindhoven, Holland)
Separaats
Philips Research Reports (Eindhoven, Holland)
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Philosophical Magazine (London)
Physica (The Hague)
Physical Review (New York and Lancaster, Pa.)
Physical Society (London)
Proceedings
Reports of the Progress of Physics
Physical Society of Japan (Tokyo)
Journal
Plating (Jenkintown, Pa.)
Plating and Finishing Guidebook (New York)
Post Office Electrical Engineers' Journal (Epsom, Surrey)
Powder Metallurgy Bulletin (Yonkers, N.Y.)
Power and Works Engineering (London)
Pratique du Soudage (Brussels)
Product Engineering (New York)
Production Engineering Research Association (Melton Mowbray)
Bulletin
Product Finishing (London)
Products Finishing (Cincinnatti)
Pro-Metal (Berne, Switzerland)
Przeglad Techniczny (Warsaw and Lodz)
Purdue University, Engineering Experimental Department (Lafayette, Ind.)
Engineering Extension Series
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Quarterly Journal of Mechanics and Applied Mathematics (Oxford)
Queensland Department of Mines (Brisbane)
Annual Reports
Queensland Mining Journal (Brisbane)

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Revista de Química Industrial (Rio de Janeiro)
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Mitteilungen
Werkzeitung
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Journal
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Journal
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Journal and Proceedings
Royal Canadian Institute (Toronto)
Proceedings
Transactions
Royal Institute of British Architects (London)
Journal
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Journal and Proceedings
Lectures
Royal Institution (London)
Proceedings
Royal Microscopical Society (London)
Journal
Royal Mint (London)
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Royal Society (London)
Philosophical Transactions
Proceedings Section A
Royal Society of Arts (London)
Journal
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Transactions
Royal Society of Edinburgh (Edinburgh)
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Journal and Proceedings
Royal Statistical Society (London)
Journal
Rozpravy (Prague)

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Journal
Reports
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Shipbuilder (Newcastle-on-Tyne)
Shipping World (London)
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Annual Report
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Boletin Minero
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Boletin
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Reports
Société des Ingeniéurs Civils de France (Paris)
Bulletin
Mémoires

ABSTRACTS

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Proceedings
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Sif Tips
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Svarovani (Prague)
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Svetsen (Stockholm)
Swansea and District Metallurgical Society (Gorseinon)
Papers

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Technique et Science Aëronautiques (Paris)
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Teknisk Ukeblad (Oslo)
Tidsskrift for Kjemi, Bergvesen og Metallurgi (Oslo)
Times Review of Industry (London)
Tin (London)
Tin Printer and Box Maker/Canning Industry (London)
Tin Research Institute (Greenford, Mddx.)
Publications
Tin and Its Uses
Tisco Review (Calcutta)
Trieste Universita, Facolta di Scienze e d'Ingenieria (Trieste)
Publicazioni

United Engineering and Foundry Company (Pittsburgh)
United Effort
Universidad Nacional de La Plata. Facultad de Ciencias Químicas (La Plata, Argentine)
Contribucions
Upsala University (Upsala)
Bulletin of Geological Institute
Usine Nouvelle (Paris)

Värmländska Bergsmannaföreningens Annaler (Filipstad, Sweden)
Verein deutscher Ingenieure (Düsseldorf)
Zeitschrift
Verkstäderna (Stockholm)

Wagner Free Institute of Science (Philadelphia)
Bulletin
Waste Trade World (London)
Welder (Waltham Cross, Herts.)
Welding (London)
Welding Journal (New York)
Welding Research (London)
Werkstatt und Betrieb (Munich)
West of Scotland Iron and Steel Institute (Glasgow)
Journal
Wire Industry (London)
Wire and Wire Products (Stamford, Conn.)
Wisconsin University, Engineering Experiment Station (Madison, Wis.)
Bulletin
Woman Engineer (London)
World Engineering Conference (London)
Bulletin

Zavodskaya Laboratoriya (Moscow)
Zeitschrift für Erzbergbau und Metallhüttenwesen (Stuttgart)
Zeitschrift für Metallkunde (Stuttgart)
Zinc Alloy Die Casters Association (Oxford)
Miscellaneous Publications
Zinc Bulletin (Oxford)
Zinc Development Association (Oxford)
Abstracts

ABSTRACTS

ORES—MINING AND TREATMENT

Britain Gets Half Its Iron from Its Lean Ores. D. D. Howat. (Engineering and Mining Journal, 1949, vol. 150, June, pp. 66–69). Difficulties encountered in the preparation and smelting of lean Northampton ores and the crushing, screening, bedding, calcining, and sintering of the ores are briefly reviewed.—R. A. R.

Iron Ore Sintering. H. U. Ross. (Canadian Mining Journal, 1949, Apr., pp. 79–84). The principles of sintering are explained and the Greenawalt and Dwight-Lloyd sintering machines and processes are described and their merits compared.—R. A. R.

Concentration Scheme of Pervouralsk Titano-Magnetites. N. G. Tyurenkov. (Gornyĭ Zhurnal, 1948, vol. 122, No. 7, pp. 36–38 : Chemical Abstracts, 1949, vol. 43, May 10, cols. 3325–3326). The mineralization of this deposit is very irregular. Besides rich ores, the deposit contains poor ores requiring fine milling prior to concentration. Five schemes for processing this ore are discussed. The preferred scheme combines the best features of cost and degree of extraction. According to it, the parent ore is crushed to 25 mm. size and subjected to dry magnetic separation which yields a concentrate and middlings (I). The former is screened to yield a fraction > 8 mm. which forms the lump concentrate and a fraction < 8 mm. (II). I is subjected to a second dry magnetic separation to yield tailings (discard) and middlings. The latter is combined with II, ground to 65 mesh, and subjected to wet magnetic separation. This operation yields tailings (discard) and a fine concentrate. The quantity of iron in the product is 94·6% of its quantity in the ore.

Magnetizing Roast of Iron Ores. A. M. Parfenov. (Gornyĭ Zhurnal, 1948, vol. 122, No. 9, pp. 29–32 : Chemical Abstracts, 1949, vol. 43, May 10, col. 3326). A discussion of the transformation of Fe_2O_3 into Fe_3O_4 in relation to temperature and reducing gas (carbon monoxide) concentration, is presented.

Iron Ore Beneficiation. F. X. Tartaron. (Mining Engineering, 1949, vol. 1, May, Section 1, pp. 14–18). As reserves of high-grade iron ore diminish in America, increased interest is being shown in the beneficiation of low-grade deposits. An account is presented of different methods of beneficiating these ores and the areas in which they are employed.—J. C. R.

FUEL—PREPARATION, PROPERTIES, AND USES

The Costing of Heat and Power. Paper 1—Fundamentals of Costing. B. C. Newbury. (Journal of the Institute of Fuel, 1949, vol. 22, Apr., pp. 177–180). An attempt is made to present a picture of the costing department for a steam-raising plant, and recommendations are made on selecting a system of costing, heat balancing and metering, and calculating procedure.—R. A. R.

The Costing of Heat and Power. Paper 2—Analysis of Overloads and Differential Costs. G. W. Pirie. (Journal of the Institute of Fuel, 1949, vol. 22, Apr., pp. 180–181).

The Costing of Heat and Power. Paper 3—Allocation of Boiler Plant Costs on the Basis of the Available Energy Content of the Steam. F. Booley. (Journal of the Institute of Fuel, 1949, vol. 22, Apr., pp. 181–183).

The Costing of Heat and Power. Paper 4—Using Information Derived from Costing Heat and Power. J. R. Rylands and A. B. Tillotson. (Journal of the Institute of Fuel, 1949, vol. 22, Apr., pp. 183–188).

Flame Research and the Practical Man. (Coke and Gas, 1949, vol. 11, May, pp. 166–168). Papers on the gas flame issued by the Gas Research Board are reviewed. Kilham showed in 1947 that the amount of heat transferred to a solid rod by a carbon monoxide flame was approximately that expected from a stream of inert gases at the same temperature, but now, using hydrogen, he finds that, under some conditions, considerably more heat than this is imparted to the solid. In the case of a hydrogen-air flame playing on a piece of alumina, the increase in heat transfer is sometimes as much as 30%. Townend referred particularly to a special kind of energy transfer occurring at temperatures just under red heat to which the name 'cando-luminescence' has sometimes been applied. Scholefield and Garside have studied the behaviour of pinhole flames as the pressure is increased.
R. A. R.

The Metering of Liquid Fuel Flow.—Laboratory Tests on Metering Pumps for Heavy Fuel Oil. J. Stringer and A. F. Gay. (Journal of The Iron and Steel Institute, 1949, vol. 163, Sept., pp. 71–80). Metering pumps are being investigated as a means of overcoming some of the difficulties of liquid flow measurement and control. This report contains results of laboratory tests on several types of reciprocating pump, and some information on rotary pumps. A high standard of accuracy is obtained from variable-stroke swash-plate-driven reciprocating pumps made to close limits of dimensional accuracy and with driven valves. An outline is given of proposed intensive trials in a steelworks to determine the rate of wear of such a pump under service conditions and its effect on metering accuracy.

The 'Ring Balance' Flow Meter. S. J. G. Taylor. (Metallurgia, 1949, vol. 39, Apr., pp. 305–308). The 'ring balance' flow meter for fuel gases consists essentially of a hollow round ring with a blanking plate inserted at the top and the lower section containing a liquid, usually paraffin. The two sides of the blanking plate are connected by flexible tubes to a connecting head from which fixed tubes run to the high and low pressure sides, respectively, of an orifice plate, or other pressure differential producer, in the fuel gas main where it is desired to measure the flow. The ring is pivoted at its centre of gravity and any pressure difference displaces the liquid in it. A suitable pointer or pen and chart indicate the movement of the ring. The theory of this meter is explained, and its practical aspects and possible sources of error are pointed out.—R. A. R.

Indirect Reheating of Blast-Furnace Gas by Boiler's Waste Gases. P. Pourchot. (Chaleur et Industrie, 1949, vol. 30, May, pp. 105–111). This paper is a mathematical discussion of the advantages obtainable by applying waste heat from boiler-flue gases to preheat blast-furnace gas used for firing the boilers.—R. F. F.

The Swelling of Bituminous Coals. M. Ussar. (Berg- und Hüttenmännische Monatshefte der Montanistischen Hochschule in Leoben, 1949, vol. 94, May, pp. 103–108). The characteristics of a number of German coals and their effects on the swelling properties were investigated. The swelling takes place during heating to about 400° C. and it is caused by the penetration of bitumen (formed partly by the depolymerization of 'bitumen formers') penetrating into the 'micelles' of the coal where it solvatizes (causes the association of molecules of solvent with solute ions or molecules). Coal must contain a certain amount of bitumen to produce swelling. If there is too little it volatilizes too quickly. All the petrographic structural components contribute with the bitumen to the swelling, but the total swelling pressure cannot be determined from the arithmetic mean of the partial swelling pressures of the individual components.
R. A. R.

A Physical Approach to the Theory of Coking. N. Berkowitz. (Fuel, 1949, vol. 28, May, pp. 97–102). The results of experiments are cited to show that the ability of coal to form coke is not due to the presence of a fusible component in the coal. It is suggested that a satisfactory explanation of coking can be obtained from a consideration of the physical structure of bituminous coals, and that softening and thermal swelling are due to mobility of 'micelles' and to the pore structure of coals respectively. The actual chemical constitution of coal may, from the standpoint of a coking theory, be very much less important than has hitherto been supposed.
R. A. R.

Apparatus for the Carbonization Assay of Small Samples of Coal and Oil-Shale. T. E. Dancy and N. J. de Jersey. (Fuel, 1949, vol. 28, May, pp. 109–113). An apparatus is described in which carbonization assays upon samples of the order of 2·5 g. may be conducted at 600° C. The results for three Australian coals are given and compared with those obtained by the Gray-King method.—R. A. R.

Some Factors Influencing the Life of Silica Coke Oven Batteries. H. Kerr and A. Taylor. (Yearbook of the Coke Oven Managers' Association, 1949, pp. 235–258). Some forms of damage to coke ovens become apparent in the first few years, but more usually the general occurrence of damage increases from the fourth to the twelfth year of working. The damage which finally shuts down a battery is usually deterioration of the oven walls. Four types of wall damage are dealt with at length; these are: (1) failure of the oven ends; (2) failure due to high temperatures and fusible coal ash; (3) failure due to erosion with wet coal; and (4) mechanical spalling of headers. Once an oven has been built, the condition of the coal charged and regular heating and operation are the essential factors governing the life of the oven brickwork.—R. A. R.

The Replacement Rate for Coke Ovens. R. J. Barritt. (Yearbook of the Coke Oven Managers' Association, 1949, pp. 145–156). The author attempts to assess the probable effect on the coke-making industry of the reduced supply of silica and semi-silica bricks. In 1934–37 the makers could supply sufficient bricks to build about 370 ovens a year, whereas the present rate of supply is at about half this rate. At the present rate of supply and building, existing batteries will have failed before the necessary replacements have been built.—R. A. R.

The Dry Cooling of Coke. (Engineering, 1949, vol. 168, Aug. 26, p. 206). The Ministry of Fuel and Power sponsored a 'Working Group' to report on the economic advantages, if any, of employing dry coke-cooling processes, and this article summarizes the report. Dry-cooled coke breaks down on handling into somewhat smaller pieces than does wet-quenched coke, but there is apparently very little more breeze; the resulting coke is harder, does not break down in the blast-furnace and is claimed to show much lower fuel consumption than wet-quenched coke under equal conditions. There is no certainty whether the metallurgical value of dry-cooled coke is higher than that of the wet-quenched product.
R. A. R.

Technological Education in the Coke Oven Industry. R. J. Sarjant. (Yearbook of the Coke Oven Managers' Association, 1949, pp. 157–169).

Liquefaction for Separating Hydrogen from Coke-Oven Gas. P. Guillaumeron. (Chemical Engineering, 1949, vol. 56, July, pp. 105–110). The 'Air Liquide' and the German processes of producing hydrogen from coke-oven gas are described.
R. A. R.

Steam-Jet Blowers in Gas-Producer Operation. G. Neumann. (Stahl und Eisen, 1949, vol. 69, May 12, pp. 328–329). When steam-jet blowers were used to supply the combustion air to gas producers before the time of motor-driven blowers all the mechanical energy of the steam was lost at the control valve before the steam entered the blast pipe. The author explains the theory of a design in which the steam jet passes through a nozzle of appropriate shape in series with the air

fan so that the mechanical energy is utilized and the overall efficiency is greatly improved.—R. A. R.

Gas Purification—Wet and Dry. J. R. Marshall. (Yearbook of the Coke Oven Managers' Association, 1949, pp. 219–234). The history and development of wet and dry processes for cleaning gas, in particular the removal of H_2S and HCN, are reviewed.—R. A. R.

TEMPERATURE MEASUREMENT AND CONTROL

Dynamic Impedance and Sensitivity of Radiation Thermocouples. P. B. Fellgett. (Proceedings of the Physical Society, Section B, 1949, vol. 62, June 1, pp. 351–359). In the case of a radiation thermocouple, or thermopile, the effect of a current passing through it is to heat or cool the junctions by the Peltier effect; thermo-electric electromotive forces are produced, opposing the flow of current, and this effect, the author shows, is proportional to current. It is therefore equivalent to a dynamic impedance added to the ohmic resistance of the thermocouple. Methods of measuring this dynamic impedance are given, and it is shown that values for the thermo-electric power, the heat loss in the receiver of the thermocouple, and its time constant can be deduced. The absolute sensitivity of the thermocouple may also be determined, and comparisons are made with previously established sensitivity formulæ.—J. P. S.

Temperature Measurement. H. Thompson and E. H. Lloyd. (Sheet Metal Industries, 1949, vol. 26, June, pp. 1259–1263). The principles of thermo-electric, radiation, and optical pyrometers are explained and some temperature-recording instruments are described.—R. A. R.

Automatic Control of Furnaces. L. Walter. (Iron and Coal Trades Review, 1949, vol. 159, July 8, pp. 91–95; July 15, pp. 209–214). Automatic control of heat-treatment furnaces includes control of roof temperature, draught, pressure, fuel/air ratio, and valve control. The mode and type of control to be used depends on controllability of a furnace which in turn is related to design and working method. The author describes how furnace characteristics influence the choice of controller. Various examples of automatic control of furnaces are given, and fuel/air ratio control is dealt with in some detail.

Notes on Temperature Regulation of Furnaces Used for Thermal Analyses. E. M. J. Mulders. (Metalen, 1949, vol. 3, Apr., pp. 165–168: Central Instituut voor Materiaal Onderzoek, Afdeling Metalen, 1949, Apr., pp. 1–4). [In Dutch]. An automatic temperature regulating device based on the e.m.f. from a thermocouple causing a mirror galvanometer to reflect a beam of light on to a photoelectric cell is described; it is designed to control an electric furnace heating melts in small crucibles.—R. A. R.

Fundamentals of Automatic Temperature Control. L. Walter. (Industrial Heating Engineer, 1948, vol. 10, May, pp. 81–83, 95; July, pp. 129–132; Sept., pp. 158–162; Nov., pp. 216–219; 1949, vol. 11, Jan., pp. 26–29; Mar, pp. 66–69). Part I of this series deals with the basic factors influencing measurement and control of temperature. In Part II the principles and the construction of temperature regulators are described. Part III discusses control problems and certain process characteristics. In Part IV proportional, floating, and proportional plus floating modes of control and their suitability for different processes are considered. Part V is concerned with control mode characteristics, the causes and cure for 'hunting' being dealt with. In Part VI notes are given on the choice of type of controller, and a proposed glossary of terms for automatic control is put forward.

R. A. R.

REFRACTORY MATERIALS

Firing Is the Most Important Step in the Manufacture of Refractories. J. O. Everhart. (Brick and Clay Record, 1948, vol. 113, No. 3, p. 64: British Ceramic Abstracts, 1949, Mar., p. 86A).

Determination of the Volume Porosity of Objects Made from Very Lean Ceramic Materials. V. V. Radin. (Zavodskaya Laboratoriya, 1949, vol. 15, June, pp. 748–749). [In Russian]. An equation is deduced, with the aid of which the apparent porosity of very lean (90% and above) ceramic material of a given composition can be calculated from known properties of the clay and grog. Some porosities calculated for two carborundum–clay mixtures are presented and shown to be only slightly lower than the observed values. The carborundum and clay contents of one mixture were 94% and 6%, respectively, the corresponding figures for the other being 90% and 10%, while the contractions on calcination of the clay in the two cases were 9·3% and 10%, respectively.

S. K.

Dolomite Brick in Metallurgy. G. V. Kukolev and D. I. Kivin. (Stal, 1947, vol. 7, pp. 539–544 [in Russian]: Chemical Abstracts, 1949, vol. 43, May 10, col. 3581). Dolomite brick with a CaO saturation coefficient of 0·85–0·90, tested in rotary cement kilns, electric steel-melting furnaces, cupolas, and open-hearth furnaces, proved to be better or at least equal to magnesite and chrome-magnesite bricks. Excessive shrinkage, deformation at high temperature, porosity and spalling of dolomite brick were eliminated by: (*a*) a minimum content of sesquioxides; (*b*) a maximum CaO saturation coefficient, but less than 1; (*c*) high MgO content; and (*d*) dense structure. The clinker for the brick was obtained by clinkering dolomite at 1480–1600°. The brick was made of ground clinker with the addition of either raw dolomite or metallurgical magnesite and fired at 1570–1590°.

High Temperature Properties of Magnesia Refractories. J. C. Hicks and B. Davies. (Iron Age, 1949, vol. 164, Aug. 11, pp. 98–105, 157–158). High-temperature tests made on high-purity magnesia obtained from sea water reveal the following facts: (1) Refractory bricks can be made from this material without the use of binding materials, and should be serviceable up to 3800° F. (2100° C.); (2) low-melting-point minor additions may migrate towards the cooler portions of the refractory; this will bring about significant changes in physical and thermal properties, concentration of the oxides of calcium, boron, and silicon in particular reducing the refractoriness-under-load; (3) small amounts of iron and aluminium oxides will not diffuse below 3700° F. (2040° C.) in a high-purity magnesia; and (4) spalling resistance is reduced after very prolonged exposure to a temperature of 3800° F. (2100° C.).

J. P. S.

The Characteristics of Spalling Cracks in Silica Bricks. I. L. Piryatinski and I. S. Schwartzmann. (Ogneupory, 1948, vol. 13, p. 211: British Ceramic Abstracts, 1949, Mar., p. 99A).

Silica Brick from the Quartzites of the Grunchbulaksk, Narchinsk and Ashiaksk Deposits. V. A. Bron. (Ogneupory, 1948, vol. 13, p. 301: British Ceramic Abstracts, 1949, Apr., p. 121A).

A Study of Silica Refractories by Torsion Methods. S. S. Das and A. L. Roberts. (Transactions of the British Ceramic Society, 1949, vol. 48, July, pp. 215-234). Torsion tests were used to study the mechanical properties at high temperatures of two commercial silica refractories (one coarse-grained, the other fine-grained). The results obtained bear out the high-load-bearing capacity of silica and point to a relationship between crystals and matrix. Silica is most liable to spall when the temperatures are below the cristobalite inversion range.—R. F. F.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Progress and Problems in Blast-Furnace Design. C. G. Hogberg. (Blast Furnace and Steel Plant, 1949, vol. 37, Apr., pp. 442–444). The author discusses the value of a high ratio of working volume to hearth area; this provides a slower rate of stock travel through the critical zone of the furnace. He also deals with the preservation of stock lines, especially in the stock-line area; disintegration of the armoured area he considers is due to erosion of the brick below this area, and the solution of this problem depends upon the co-operation of ceramic engineers, designers, and blast-furnace operators.—J. P. S.

Tata 'A' Blast-Furnace Makes Over Three Million Tons. E. J. Warren. (Blast Furnace and Steel Plant, 1949, vol. 37, June, pp. 673–675). The Tata 'A' Furnace, blown in at the end of 1939, was blown out in April 1948, after a campaign of 3056 days and a total production of 3,266,455 net tons, an average production of 2·71 tons/sq. ft. of hearth per operating day. On blowing out, the lining was in good condition, except that the stock-line protection plates had all fallen out; the original big bell was still in use, though the small bell had been changed three times.—J. P. S.

Influence of Agglomerated Ore on the Blast-Furnace Charge. C. Ricci. (Metallurgia Italiana, 1949, vol. 41, May–June,

pp. 121–127). The results of operating a blast-furnace with three types of charge are examined. These charges consisted of: (1) Cogne magnetite in lump form; (2) 50% lump magnetite, 50% magnetite, sintered with blast-furnace gas; and (3) 62·5% lump magnetite, 37·5% sintered magnetite. With the first two types of charge, only metallurgical coke was used as fuel, with the third 22·8% of La Thuile anthracite was employed. The addition of the sinter, which is richer in iron than the raw material, appeared to increase the amount of indirect reduction and reduce the coke consumption.—R. F. F.

Blast-Furnace Operation and the Effect of Coke Quality. D. Joyce. (Yearbook of the Coke Oven Managers' Association, 1949, pp. 259–279). *See* Journ. I. and S.I., 1949, vol. 161, Feb., p. 148.

Distribution of Materials in the Blast-Furnace. Part II—Compensated Charging. H. L. Saunders and R. Wild. (Journal of The Iron and Steel Institute, 1949, vol. 163, Sept., pp. 61–70). The paper deals with the extension of the work (*see* Journ. I. and S.I., 1945, No. II, pp. 259P–286P) on the distribution of materials in the blast-furnace by comparing the results obtained in the laboratory with those on a scale intermediate between the latter and full-size furnaces. Comparative results show the reliability of the models as predictors, the necessity for some scheme of compensated charging to eliminate segregation, and the application of a scheme worked out in the laboratory to an experimental blast-furnace.

Studies Relating to the Control of Sulphur in the Production of Pig Iron. T. H. Kennedy and A. W. Thornton. (Blast Furnace and Steel Plant, 1949, vol. 37, June, pp. 676–686). Owing to the exhaustion of first-class coking coals low in sulphur, and the increase in mechanical mining, the problem of eliminating sulphur has become a serious one for the blast-furnace operator. This paper studies the problem from the aspect of slag control and describes a series of experiments whereby the amounts of dolomite and limestone in the flux charge of a blast-furnace were varied from 100% limestone, through mixtures of limestone and dolomite, to 100% dolomite. Figures were obtained for sulphur in the coke, the iron, and the slag, and for lime, silica, magnesia, and alumina. It was concluded that increasing the slag basicity is better, as a means of desulphurization, than increasing its volume, and that the inclusion in the flux charge of between 5% and 8% of magnesia equivalent to 23% to 37% of dolomite has the optimum effect, increase beyond this resulting in a reduction in the sulphur-absorption properties of the slag. The author considers the slag/iron sulphur ratio a useful index in operation where the sulphur in the coke varies greatly, as it shows at once whether a change in the sulphur in the iron is due to an increase in the sulphur entering with the charge or to inefficient desulphurization in the furnace.—J. P. S.

Reducing Coke Consumption in Iron and Steel Production. M. Tigerschiöld. (United Nations Scientific Conference on the Conservation and Utilization of Resources, Lake Success, 1949, Aug. 17–Sept. 6; Iron Age, 1949, vol. 164, July 14, pp. 84–91). The coming exhaustion of Mesabi ores and of good coking coal will present many difficult problems for the American iron and steel industry. The author describes Swedish progress in the reduction of coke consumption by the use of sinter made self-fluxing by the addition of limestone, and discusses the use of oxygen in the blast-furnace, the electric pig-iron furnace, and sponge-iron processes as means of achieving an economic solution to the fuel problem. J. P. S.

DIRECT PROCESSES

Construction and Operation of the Krupp-Renn Plant at Watenstedt. D. Fastje. (Stahl und Eisen, 1949, vol. 69, May 12, pp. 319–325). A description is given of the layout of the Krupp-Renn plant at Watenstedt which consists of ore-crushing and fuel preparation machinery, three rotary kilns 4·2 m. in dia. × 70 m. long, and crushing and screening equipment for the product. Each rotary kiln can deal with 500–600 tons of ore per day. The discharge end of the kilns, where slag attack was greatest, was lined with silica bricks (SiO_2 75–76%, Al_2O_3 20–21%). The economic aspects of production are examined. The interest on capital for the Watenstedt plant is particularly high for special reasons. It is estimated that the cost of producing pig iron by the Krupp-Renn and subsequent blast-furnace processes is about 80 RM./ton which is from 15 to 40 RM. cheaper than it would be when smelting the same Salzgitter ore in the blast-furnace alone, but in this comparison the capital charges on the Krupp-Renn plant are not taken into account; if this were done the costs of the two processes would be approximately the same.—R. A. R.

Commercial Production of Sponge Iron. P. E. Cavanagh. (Iron Age, 1949, vol. 163, June 2, pp. 67–71, 82). The author discusses the Wiberg-Söderfors process for the production of sponge iron in relation to its applicability to Canadian conditions and Canadian Steep Rock ore. A trial run at Söderfors on a quantity of this high quality, readily reducible ore is described, and it is concluded that the process would economically produce sponge iron for steelmaking, at many locations in Canada; there might also be advantages in the United States where high quality steels are required.—J. P. S.

The Treatment of Pyrites Cinder for Pig-Iron Production. H. E. Engemann. (Echo des Mines, 1948, Oct., pp. 198–199, Nov., pp. 219–220, Dec., p. 239; 1949, Jan., p. 11, Feb. pp. 32–33). After studying the difficulties preventing the the greater use of pyrites cinder and describing the German processes of Wissen and Stuerzelberg, the author suggests a new process for the recovery of sulphur and iron by the direct reduction of pyrites without forming pyrites cinder. The economic advantages of this process are calculated.

Reduction of Iron Ores and Agglomerates. L. M. Tsylev. (Bulletin de l'Académie des Sciences, U.R.S.S., Classe des Sciences Techniques, 1948, May, pp. 673–680 [in Russian]: [Abstract] Metals Review, 1949, vol. 22, Jan., p. 23). Attempts are described to solve the problem of the different reducibilities of several types of iron ore (magnetite, hematite, limonite) by partially reducing them in a hydrogen atmosphere at different temperatures and by investigating the resulting agglomerates. On the basis of microscopic examination five different schemes for the reduction process are proposed, depending on composition and structure of the treated ores.

PRODUCTION OF STEEL

Presidential Address. E. A. Davies. (Swansea and District Metallurgical Society, Oct. 17, 1947). After reviewing the position of the Society the author describes the plant of the Cwmfelin Steelworks of Richard Thomas and Baldwins, Ltd. and the conversion at these works of the acid furnaces to basic, and to the making of rimming steel for the strip mill at Ebbw Vale.—R. A. R.

Improving Bessemer Steel Quality for the User. H. A. Dickie. (Sheet and Strip Metal Users' Technical Association: Sheet Metal Industries, 1949, vol. 26, May, pp. 956–959, 970). The steelmaking practice developed at the Corby works of Stewarts and Lloyds, Ltd. is described. The nitrogen and phosphorus contents are mainly responsible for the work-hardening and embrittlement of Bessemer steel. The converter linings are made of brick so as to be thinner than the tamped monolithic lining; this means that a shallower bath of iron can be charged without loss of production, the object being to shorten the path of the air blown through the metal. Mill-scale is added to the converter in place of the usual additions of steel scrap to cause direct oxidation of the charge and prevent the oxygen in the air-blast from being used up. This practice reduces both the nitrogen and the phosphorus contents to about half the values obtained in ordinary Bessemer practice.—R. A. R.

Modern Auxiliary Steelworks Equipment. A. C. Hobson. (Iron and Coal Trades Review, 1949, vol. 159, July 29, pp. 305–312). A review is presented of developments in the design and operation of open-hearth furnace reversing valves, doors and frames, dolomite fettling machines, and fuel-oil atomizers.—J. C. R.

Chemistry Performance Survey of Open-Hearth Alloy Steels. (American Iron and Steel Institute, 1949, Contributions to the Metallurgy of Steel No. 28). Statistical analyses are made of the relation between the actual and the intended compositions of steel produced using ladle analyses of 13,743 open-hearth heats reported by the 22 plants which participated in the survey, the object being to determine the probability of the average melter foreman being able to obtain the exact content of carbon or other element at which he aims, as well as the probability that the chemical composition achieved will deviate 0·01, 0·02, 0·03% or more from this aim for a particular plant.—R. A. R.

Application of Automatic Control to Open-Hearth Furnaces. E. Whitehead. (Journal of The Iron and Steel Institute, 1949, vol. 163, Sept., pp. 1–8). The paper deals with some of the factors which require automatic control, and indicates why priority is given to air/fuel ratio, furnace pressure, roof temperature, and reversal. Attention is drawn to some of the major sources of error in the measurements, and to other difficulties which act against satisfactory control. Accuracy of control is discussed, with particular reference to the "uncontrolled variable." Finally, the basic principles of automatic control are presented in a simple non-mathematical way to enable the user to develop a sound understanding of the operation of controllers. This presentation is not intended to replace the highly specialized knowledge of the control designer, but to allow the user to co-operate with him by supplying accurate information on the characteristics of the furnace and process.

A Statistical Analysis of the Output of an Open-Hearth Furnace. F. L. Robertson and M. W. Thring. (Journal of The Iron and Steel Institute, 1949, vol. 163, Sept., pp. 31–50). Measurements of nearly all the important variables which might affect the output of an open-hearth furnace have been made on about 47 casts from a single furnace. The measurements have been analysed to determine those which are clearly related. The output is governed by the number of tons produced and the total time. The latter is broken down into charging plus melting times ($t_1 + t_2$), and decarbonizing time (t_3). It is shown that t_3 is almost entirely predictable when the carbon content of the bath at junction C_J (junction is the time when the temperature of the metal bath reaches a value which is related to its carbon content) is known. The quantity most strongly connected to $t_1 + t_2$ is the mean rate at which oil can be burned in the furnace during this period. Because C_J increases if the period $t_1 + t_2$ is decreased by a greater oil input rate, the quantity most strongly related to the total time $t_1 + t_2 + t_3$, and so the output, is the rate of oxidation of the metal during $t_1 + t_2$. The rate of oxidation is shown to affect C_J mainly by its effect on the weight of FeO in the slag at junction.

Comparison of Various Types of Charge Ore and Sinter on Production and Fuel Rate in Open Hearth Furnaces. (American Iron and Steel Institute, 1949, Contributions to the Metallurgy of Steel No. 26). Forty-six melting shops submitted replies to a questionnaire on open-hearth charging practice, production and fuel rate. The replies are analysed in 47 pages of tables.—R. A. R.

Use of Oxygen or Compressed Air for Open Hearth Combustion and Decarbonization. (American Iron and Steel Institute, 1949, Contributions to the Metallurgy of Steel No. 27). Twenty-two open-hearth steelplants replied to a questionnaire on the use of oxygen or compressed air, the method by which it is applied, and the effects on fuel consumption, life of refractories, and production. The answers are analysed and presented in 49 pages of tables.—R. A. R.

Oxygen as a Combustion Factor in the Open-Hearth Furnace. D. D. Howat. (Iron and Coal Trades Review, 1949, vol. 158, June 10, pp. 1265–1272). *See* Journ. I. and S.I., 1949, vol. 162, Aug., p. 480.

Increasing Open Hearth Production by Use of Oxygen, Better Refractories and Control of Slag. E. G. Hill. (Blast Furnace and Steel Plant, 1949, vol. 37, June, pp. 667–672, 675). The author discusses the possible increases in production attainable by the use of oxygen for combustion and for decarburization, by the use of better refractories, and especially by the employment of basic roofs and ends, and by slag control, notably in the early stages of the heat. He concludes that a total increase of 30% production, with a furnace of medium age and of all silica brick construction, is possible, given certain changes in auxiliary equipment and the acquisition of sufficient skill in the new practices: of this increase, 15% would be due to the use of oxygen for combustion, 8% to its use for deoxidation, 5% to the employment of better refractories, and 2% to charge and slag control.

J. P. S.

Distribution of Oxygen between the Metal and the Basic Slag from the Point of View of the Ion Theory. O. Esin and V. Kozheurov. (Journal of Applied Chemistry, U.S.S.R., 1948, vol. 21, pp. 765–774 [in Russian]: Chemical Abstracts, 1949, vol. 43, Apr. 25, cols. 2908–2909).

Some Statistical Observations on the Oxygen Content during Steelmaking. A. Antonioli. (Metallurgia Italiana, 1949, vol. 41, Mar.–Apr., pp. 71–79). Two fundamental hypotheses have been formulated on the mechanism of carbon reaction during steelmaking. In the first, the carbon–oxygen reaction in the homogeneous phase is considered to be the most important stage of the boiling period. In the second, the exchange processes between the heterogeneous phases are thought to be the more significant. After examining the relations between oxygen and other variables on the basis of both hypotheses an important group of results is considered statistically. As the determination of oxygen is uncertain, no definite conclusions could be drawn but it is pointed out that the results obtained agree with the hypothesis of the reaction between the heterogeneous phases, but not with that of the reaction in the homogeneous phase.—R. F. F.

Carburization of Molten Steel for Castings. N. I. Vinogradov. (Stal, 1947, vol. 7, pp. 537–539 [in Russian]: Chemical Abstracts, 1949, vol. 43, Apr. 25, col. 2910). For steel castings use is frequently made of steel made in basic open-hearth furnaces. Steel containing 0·08–0·13% of carbon, trace silicon, and 0·30–0·60% of manganese is too soft. Ordinarily the furnace is tapped into a large ladle and the melt transferred to a preheated smaller ladle for casting. The steel is carburized by adding 1–2 kg. of graphite per ton of metal. Graphite was added to the metal as it flowed from the larger into the smaller ladle. In addition to graphite, 5 kg. ferromanganese, 3 kg. ferrosilicon, and 0·4 kg. aluminium per ton of metal, were added.

FOUNDRY PRACTICE

The Cupola. M. Stap. (Metalen, 1949, vol. 3, May, pp. 196–205). [In Dutch]. The cupola is described and the theory and practice of its operation are explained.—R. A. R.

Quality Control in the Foundry. R. C. Byers and J. Dickman. (Foundry, 1949, vol. 77, May, pp. 102–105, 272, 274). The importance of limiting the range of variations in carbon and silicon contents in iron castings is discussed. An account is given of attempts to control the hardness of small bell-shaped castings for the ends of motors, so as to avoid the necessity of annealing them prior to machining. Data indicated that there was a significant difference in silicon content between machinable and unmachinable castings, and the relationship between silicon content and Brinell hardness was accordingly examined. It was found that for each 0·1% of silicon the hardness was lowered by 8 to 10 points on the Brinell 'B' scale.—J. C. R.

Conversion of Steel Scrap into Foundry Pig Iron in the Cupola. H. Reininger. (Die Technik, 1949, vol. 4, Apr., pp. 143–149). German experience in the melting down of steel scrap in the cupola to make foundry pig iron is reported. The absorption of silicon is much greater when it is added in the form of 'packets' in which the silicon is surrounded by a flux, than when added as ferrosilicon. A charge of thick lumpy pieces of scrap takes up less carbon than when the scrap has a greater surface/thickness ratio. Suitable equipment consists of a cupola at least 1 m. in inside dia. and a tilting forehearth into which the iron passes *via* a slag trap and runner.—R. A. R.

Graphite Resistor Furnace Melting Practice. B. N. Ames and N. A. Kahn. (American Foundrymen's Society, 1949, Preprint No. 52). A graphite rod electric resistance furnace capable of melting 300 lb. of steel is described. The resistors consist of two pairs of $1\frac{3}{4}$ in. graphite rods meeting in the centre of the furnace. Two-phase current is supplied from a transformer because the furnace is intended for use on ships. A first heat of 300 lb. of steel consumes about 2300 kWh./ton, but for the third and fourth consecutive heats this is reduced to just over 1100 kWh.—R. A. R.

Contribution to the Study of Controlled Graphitization in Cast Irons. F. Gaty. (Revue Universelle des Mines, 1949, Series 9, vol. 5, July, pp. 231–237). After a review of the technique of inoculation of cast iron, the author reports a series of tests on inoculation carried out at the metallurgy laboratories of the University of Liége. From the results obtained general rules, which must be followed in industrial practice, are suggested.—R. F. F.

Progress Report on Nodular Iron. C. O. Burgess. (Foundry, 1949, vol. 77, May, pp. 112–115, 218, 220–225, 228, 230). The author discusses developments in research work on nodular irons. Two conditions appeared to be essential for

the conversion of flake to nodular graphite : (1) The sulphur content of the iron must be reduced to 0·02% either by using a heat initially low in sulphur or adding excessive amounts of cerium or magnesium ; and (2) the added element appears in every case to suppress graphite formation at normal temperature, that is, it behaves as a carbide-stabilizing element. He gives details of his own experiments with cerium-treated irons and of the results he obtained. An account is included of the use of magnesium and alloys with a maximum content of 30% magnesium as a nodulizing element. It has the advantage that lower carbon irons with sulphur contents up to 0·09% can be treated, but the danger of the explosive nature of magnesium and magnesium-rich alloys is stressed. Possible commercial applications of nodular irons are suggested and factors requiring clarification are noted.—J. C. R.

Stress Analysis for the Steel Foundry. R. G. Waite. (Foundry, 1949, vol. 77, May, pp. 106–107, 252, 254). An account is presented of a study of the nature and magnitude of residual stresses which are induced when plates or sections are welded in openings in castings for the purpose of making a solid wall or pressure-tight and leak-proof compartment. Such openings are necessary in foundry practice to maintain wall thickness or core positions to produce internal section members of intricate castings.—J. C. R.

Electrolytic Manganese in Acid Electric Steel Foundry Practice. F. Sillers, jun. (United States Bureau of Mines, June, 1949, Report of Investigations No. 4482). Tests were made at a steel foundry making large-link chains with cast 0·25% carbon steel produced in acid electric furnaces to determine whether electrolytic manganese could be used in place of ferromanganese. Electrolytic manganese can be used successfully in acid practice. Although the castings were prone to retain slightly more hydrogen than castings made with ferromanganese, the hydrogen was still within the permissible range and it produced no deleterious effects on the finished castings.—R. A. R.

Hungarian Sands and their Utilization in the Foundry Industry. J. Tömösközy. (Bányászati és Kohászati Lapok, 1949, vol. 4, June, pp. 246–253). [In Hungarian]. The present output of 105,000 metric tons per annum of ferrous castings in Hungary requires at least 100,000 tons of sand. The properties of the sands available and means of increasing the production are discussed.—R. A. R.

Investigation of the Control of Moulding Sand. M. Stap. (Metalen, 1949, vol. 3, Apr., pp. 175–181 : Centraal Instituut voor Materiaal Onderzoek, Afdeling Metalen, 1949, Apr., pp. 5–10). [In Dutch].

The Cement-Sand Moulding and Core Process—Introduction to Its Theory and Examples of Its Practical Application. M. Beilhack. (Neue Giesserei, 1949, vol. 36, May, pp. 138–142). The author gives the composition and describes the preparation of cement sand for moulds and cores used at the foundry with which he is connected, and describes several examples of the castings produced by this technique. The examples include a band-saw machine stand weighing about 350 kg., and a pipe junction weighing 720 kg.—R. A. R.

Handling Materials in the Coreroom. R. H. Herrmann. (Foundry, 1949, vol. 77, June, pp. 76–81, 190, 192, 193). Illustrated descriptions are given of equipment and machinery used in the mechanization of the core-room to facilitate and speed up the handling of materials.—J. C. R.

Core Dipping and Spinning. A. H. Allen. (Foundry, 1949, vol. 77, May, pp. 108–111). An illustrated account is given of improvements in the practice for the assembly and dipping of jacket and barrel cores for engine block castings for Ford V-8 power units, at the Rouge plant of the Ford Motor Co.—J. C. R.

Core Sand Practice in the Malleable Iron Foundry. E. Welander. (Foundry, 1949, vol. 77, May, pp. 120–122, 262, 264). The properties of foundry sands and factors affecting their choice and use in core-room practice are discussed.

J. C. R.

Testing the Wear Resistance of Cores. F. Roll. (Die Technik, 1949, vol. 4, Apr., p. 150). A simple apparatus for testing cores is described. In this test a core (50 mm. in dia. × 50 mm.) long is revolved at 60 r.p.m. in a chamber while a stream of granulated steel scrap falls on it from a height of 20 cm. The loss in weight of the core after 1·5 kg. of steel granules have fallen on it is measured.—R. A. R.

Solidification, Contraction and Feeding Phenomena in the Manufacture of Castings. Part IV. D. Iturrioz. (Metalurgia y Electricidad, 1949, vol. 13, July, pp. 22–25). [In Spanish]. Previous articles dealt with funnel contraction. The other main types of contraction, *i.e.*, interdendritic and micro-contraction, are described here. The mechanism of crystallization is first explained. Complications arise from differences in size and disposition of grains, and three cases are discussed : (*a*) Continuous envelopment where all crystals are rigidly united ; (*b*) skeleton envelopment, where crystals are rigidly united but contact is at points only so that liquid can penetrate the interstices ; and (*c*) brittle type, where the crystals are not in contact with each other but are suspended in the liquid. Finally, the influence of shrinkage cavities on the mechanical properties is discussed. This results in reduction of section and stress concentration ; however, spherical pores have a lesser effect.—R. S.

HEATING FURNACES AND SOAKING PITS

Heat Transfer in Metallurgical Furnaces. C. Hulse and R. J. Sarjant. (Société Française de Métallurgie, Revue de Métallurgie, Mémoires, 1949, vol. 46, May, pp. 297–308). The authors describe existing methods available for the study of the variable flow of heat in solids and give details of new mechanical methods of calculation. These methods show clearly how important it is to achieve uniform heating, by which the time required to carry out a given heating procedure can be much reduced.—A. E. C.

New Slab Heating Furnace Uses Coke Oven Gas. (Gas Heat in Industry, 1948, Dec., pp. 69–71). A new slab reheating furnace at the Gary Works of the Carnegie-Illinois Steel Corp., is described. The inside dimensions of the furnace are 21 ft. wide × 80 ft. long. It has separately controlled heating and soaking zones. The lining is of silicon–carbon-type bricks and the equipment includes a recuperator with silicon–carbide tubes.—R. A. R.

A New Tube Reheating Furnace. E. S. Kopecki. (Coke and Gas, 1949, vol. 11, Mar., pp. 91–95). A description and some operating data are given respecting the high-speed continuous gas-fired furnace for tubes developed by the Babcock and Wilcox Tube Co., Beaver Falls, Pa. The furnace is for reheating the tubes to 2000–2200° F. before the final sizing operation. The consumption of natural gas has been reduced from 2662 cu. ft. in the former furnace to 1051 cu. ft./ton in the new furnace. (*See* Journ. I. and S.I., 1947, vol. 157, Sept., p. 144).—R. A. R.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Compact, Versatile Heat Treating Shop of Cincinnati Shaper Co. (Industrial Heating, 1949, vol. 16, May, pp. 808–816, 822). The heat-treating equipment and processes at the Cincinnati Shaper Co., Cincinnati, Ohio, are described and illustrated. Gears, castings, and flame-cut parts of framework represent the types of material treated.—R. A. R.

Some Recent Heat Treatment Furnace Installations. (Metallurgia, 1949, vol. 40, May, pp. 44–53). The modern heat-treatment furnaces which are described and illustrated include (1) a walking-beam furnace for non-ferrous slabs, (2) an oil-fired furnace for aluminium billets, (3) an oil-fired stress-relieving furnace with inside dimensions of 16 ft. 6 in. × 16 ft. 6 in. × 27 ft. 6 in. long, for pressure vessels and boilers, (4) a carburizing furnace applying the Homocarb method, (5) equipment for carburizing with prepared town's gas, which incorporates heat-resisting removable retorts ; these retorts are transferred to controlled cooling pits after the actual gas carburizing, (6) a town's gas fired radiant tube cell furnace in which lithium is introduced into the working chamber, (7) salt-bath installations with automatic conveyors for handling small parts, (8) a continuous heat-treatment plant, electrically heated, for 3 tons of steel forgings per hour, (9) the Drever gas-tight continuous strip annealing plant (10) malleablizing furnaces, (11) tool-hardening furnaces, and (12) induction heating equipment.—R. A. R.

Localized Heat Treatment. W. F. Sorenson. (Iron Age, 1949, vol. 164, July 28, pp. 52–57). A rotary-flame heater for small parts, constructed by the Yale and Towne Manufacturing Co., Philadelphia, is described and illustrated.

Liquid Salt Baths. J. A. McElgin. (Canadian Metals and Metallurgical Industries, 1949, vol. 12, Apr., pp. 14, 15, 37–42; May, pp. 18–20, 29, 33, 35–38). Liquid salt baths, their advantages and application for a wide variety of heat-treatment are discussed.—R. A. R.

Production Induction Heating. W. E. Benninghoff. (Industrial Heating, 1949, vol. 16, May, pp. 792–804, 822). Several applications of induction heating for heat-treating bars and parts of various shape, brazing and soldering, and for assembling shrink fits, are described and the advantages of the process are pointed out.—R. A. R.

How to Handle Parts in Induction Heating. C. E. Glick and R. E. Harlepp. (Machinist, 1949, vol. 93, July 23, pp. 460–464). The handling equipment in use with induction furnaces is described and illustrated and the materials used in the construction of this type of equipment are discussed.
R. F. F.

Some Applications of the High Frequency Induction Heating Process. J. C. Howard. (Metallurgia, 1949, vol. 40, May, pp. 37–43). After explaining the principles of high-frequency induction heating the author describes some modern applications such as the hardening of crankshafts and tractor links, the continuous hardening of steel bars, the heating of parts for forging, and for brazing and soldering.—R. A. R.

Isothermal Heat Treatment of Large Steel Castings. D. Rosenblatt. (Iron Age, 1949, vol. 163, June 30, pp. 42–47). Factors involved in the heat-treatment of large alloy steel castings are considered; annealing, normalizing and quenching, quenching and tempering, martempering, and isothermal treatments are described, and the advantages of the last-named are considered to be sufficient to warrant a special installation.—J. P. S.

Steam Atmosphere Used to Heat Treat and Improve Surface Properties. F. L. Spangler. (Materials and Methods, 1949, vol. 29, May, pp. 56–58). To produce a hard blue oxide coating on steel tools and parts they are tempered in a steam atmosphere. The charge in the furnace is preheated to 650–750° F. in air; the furnace is then purged with steam for about 30 min. and brought up to about 1009° F. with steam passing through at a pressure of about 4 in. of water. The parts may be air-cooled or quenched in liquid. The treatment of iron parts made by powder metallurgy with steam increases the hardness, and resistance to wear and corrosion.—R. A. R.

The Strength-Reducing Annealing in the Continuous Furnace of Low-Carbon Steel Strip Produced by Different Processes. A. Pomp, J. Brockhaus, and G. Niebch. (Mitteilungen aus dem Max-Planck-Institut für Eisenforschung: Archiv für das Eisenhüttenwesen, 1949, vol. 20, May–June, pp. 199–204). Cold-rolled low-carbon steel strip 35 × 1 mm. in section, was produced by five different processes (*a*, open-hearth furnace, rimming; *b*, basic-Bessemer, rimming; *c*, *d*, and *e*, special blowing process, rimming, silicon-killed, and aluminium-killed, respectively) in a continuous furnace 12·5 m. long at temperatures of 400°, 450°, 500°, 550°, 600°, and 700° C. The properties were compared with those obtained after annealing in a batch furnace and found to be in no way inferior. The annealing temperature in the continuous furnace had to be 100° C. higher with a maximum of about 600° C., whereas in the long-time batch process 500° C. is the maximum. Under certain conditions better elongation at the same tensile strength can be obtained in the continuous process.—R. A. R.

Conductors Stress Relieve Welds. C. C. Roberts. (Machinist, 1949, vol. 93, July 23, p. 465). The circuits for the automatic preheating and annealing of welded pipes are described. The heating units used consist of resistance wires encased in refractory insulation material.—R. F. F.

Hardening Enhances Properties of Malleable Iron. K. Rose. (Materials and Methods, 1949, vol. 29, June, pp. 56–58). Pearlitic malleable irons to A.S.T.M. specification A-220 are discussed. When a hard surface is required on a malleable iron casting it should have an initial hardness of about 197 Brinell before local hardening by a rapid treatment such as induction heating. The pearlitic malleable irons are obtained by varying the malleablizing process so as to give incomplete graphitization. In the case of a pearlitic malleable iron known as 'Z-metal' graphitization is retarded by manganese in amounts of 0·5–1·0% above that which would be used to produce a similar alloy in ordinary malleable iron.
R. A. R.

Effect of Hardening from the Heat of Rolling in the Case of Heat-Treatable Steels. F. Bollenrath and H. Kiessler. (Stahl und Eisen, 1949, vol. 69, Apr. 28, pp. 287–301). Four different techniques for hardening from the heat of rolling or forging, previously described by (*a*) R. Schäfer and W. Drechsler, (*b*) P. Drastik, W. Gatzek, and A. Schneider, (*c*) O. Kukla, W. Küntscher, and H. Sajosch, and (*d*) W. Stich, are critically examined and the hardness and other test data for steels heat-treated in this way are compared with the properties obtained by reheating and quenching in the conventional manner. There is no evidence that better properties are obtained by the former method, but in all four techniques the economies obtained by obviating the cost of reheating are very great. Present and future applications of the process are discussed.—R. A. R.

The Effect of Tempering Time on the Hardness of Tool Steel. A. Pomp and A. Krisch. (Mitteilungen aus dem Max-Planck-Institut für Eisenforschung: Archiv für das Eisenhüttenwesen, 1949, vol. 20, July–Aug., pp. 255–262). The reduction in hardness with time on tempering for up to 1000 hr. at temperatures in the 50–270° C. range was measured on 24 tool steels. These steels comprised three high-carbon steels; one each vanadium, chromium–silicon, chromium–manganese, and chromium–vanadium steels; three chromium–tungsten, two manganese–vanadium, and three chromium–nickel–molybdenum steels; and three high-speed steels. The hardness diminished at an increasing rate during the first hour in the 90–270° C. tempering range, whilst at 50° C. the hardness increased slightly. After 1000 hr. at one temperature the hardness was much less, most of the reduction occurring in the first 48 hr. The maximum rate of loss of hardness, apart from the first hour, took place on tempering at 150° C. and the minimum at 210° C. The effect of time was negligible on the high-speed steels at 270° C. All the above results relate to the tempering of previously hardened specimens. Long-time tempering at 90° C. of specimens previously tempered for 1000 hr. at 120°, 150°, and 180° C., showed that, contrary to the results with previously quenched specimens, the time had no effect on the hardness.—R. A. R.

The Cooling Capacity of Liquid Hardening Media. W. Peter. (Mitteilungen aus dem Max-Planck-Institut für Eisenforschung: Archiv für das Eisenhüttenwesen, 1949, vol. 20, July–Aug., pp. 263–274). The silver-ball technique was applied to investigate the cooling capacity of water, oils, aqueous salt solutions, and molten metals and salts. Cooling curves for gas-free distilled water and for distilled water saturated with (*a*) nitrogen, (*b*) air, (*c*) oxygen, and (*d*) carbon dioxide are presented, and the effect of hard water is shown. The cooling rates of mineral oil can be modified over a wide range by various additions and the range between the rates for water and oil can be covered by aqueous solutions of calcium chloride, magnesium chloride, and sodium hydroxide.
R. A. R.

FORGING, STAMPING, DRAWING, AND PRESSING

The Parallel Development of Heavy Self-Contained Hydraulic Presses in the United States and Great Britain. F. H. Towler. (Transactions of the American Society of Mechanical Engineers, 1949, vol. 71, July, pp. 501–514). The parallel development of heavy self-contained hydraulic presses in the United States and Great Britain is surveyed. This type of press has been made possible by the development of a compact high-speed pump using oil as the hydraulic medium; the construction of the pump has a considerable influence on the design of the press. In the United States the tendency has been to use pumps of the rotary-valve type operating at a maximum pressure of about 3000 lb./sq. in. with a comparatively thick oil of around 300 Saybolt viscosity at 100° F. In Great Britain there is a growing tendency to use pumps of the seated-valve type capable of a working pressure of 7000 lb./sq. in. and using a thin oil of 60 Saybolt viscosity at 100° F. The relative advantages and disadvantages are discussed.

Rubber Bonded to Metal Rapidly Becoming a Production Tool. W. A. Keetch. (Steel, 1949, vol. 124, May 16, pp. 82–85, 116). Rubber bonded to metal may be used to absorb the vibration of moving parts, to cushion impacts, as in the packing of fragile stores, and in floating holders for lathe tools and drills. Rubber bearings when flood-lubricated with water have many functions, especially when handling

grit-laden fluids. The article also briefly mentions the use of rubber formers in sheet metal work.—J. P. S.

The Production of Fibrous Structures in Metals during Plastic Deformation. P. Bastien. (Revue Générale de Mécanique, 1949, vol. 33, Mar., pp. 112–118). The author discusses the effect on the structure of metals (especially steels) of plastic deformation during hot and cold work. In steels, after plastic deformation, a fibrous macrostructure will result due to the dendritic segregation, a fibrous microstructure will be especially pronounced when the work has been carried out at a temperature between A_3 and A_1, and, as a result of the mechanism of slip, a fibrous structure on the crystalline scale will be formed.—R. F. F.

The Manufacture of Hoes in Brazil. J. G. Haenel. (Steel Processing, 1949, vol. 35, Apr., pp 187–190). The manufacture of one-piece hoes, in which the handle holder is integral with the blade, is described : the bulk of the article is mild steel, a higher-carbon facing being forge-welded on during the process.—J. P. S.

The Photo-Grid Process and Machine Design. G. A. Brewer. (British Journal of Photography, 1947, vol. 94, p. 271 : Metallurgical Abstracts, 1949, vol. 16, May, p. 597). The process is used in studying metal deformation, *e.g.*, in cold pressing. A rectangular or polar grid is recorded photographically on the surface of the specimen. The distortion of the grid, arising from working the metal, is used to determine the local deformations. Instructions for recording the grid are given.

Extrusions Last Longer, Cost Less Than Forgings. C. L. Stevens and G. Vennerholm. (SAE Journal, 1949, vol. 57, July, pp. 43–46). **Applications of Hot Extrusion Methods.** C. L. Stevens and G. Vennerholm. (Steel Processing, 1949, vol. 35, July, pp. 362–367). *See* Journ. I. and S.I., 1949, vol. 163, Oct., p. 226.

ROLLING-MILL PRACTICE

Power Distribution Systems and Equipment for Steel Mills. D. L. Beeman. (Blast Furnace and Steel Plant, 1949, vol. 37, June, pp. 687–696). The electric power loads in steel rolling mills are unusual in that they consist of large motors of several thousand horsepower each, concentrated in relatively few locations, and operating either at very low power, as when idling, or at high peak outputs, as when rolling metal. Great reliability is required, as breakdown means expensive production losses. The author discusses several aspects of the distribution of power under these conditions. The selection of voltage is linked with cost and flexibility, and the advantages of 6·9 and 13·8-kV. systems are discussed, together with the types of distribution network and switchgear required ; the necessary arrangements for safety in switchgear and substations are considered.—J. P. S.

Bevelling and Rounding Off in Roll Pass Design. A. E. Lendl. (Stahl und Eisen, 1949, vol. 69, Apr. 28, pp. 306–308). To prevent fin formation in the rolling of bars it may be necessary to provide passes which produce bevels on the two upper edges. In this paper formulæ for calculating the dimensions of the bevels are derived and examples of their application are given.—R. A. R.

The Mechanization of the Sheet Mill. W. Krämer. (Stahl und Eisen, 1949, vol. 69, May 26, pp. 359–371). A review is presented of the development of mechanical aids to sheet mills, with special reference to mill tables of several types including tipping, roller, and chain tables of which 21 different designs are described and illustrated. Roller transport between furnaces and mills, and between mill stands, is also described.—R. A. R.

Evolution of the Reversing Hot Strip Mill. T. W. Lippert. (Iron Age, 1949, vol. 163, May 19, pp. 86–89). The history and development of hot reversing mills for strip, of the Steckel and hot-reeling types, are described.—J. P. S.

Extensometer Continuously Records Percent of Extension on New Strip Temper Pass Mill. (Instruments, 1949, vol. 22, Mar., pp. 226, 228). A brief description is given of an electrical recording extensometer developed by the Carnegie-Illinois Steel Corporation.

LUBRICANTS AND LUBRICATION

Friction and Lubrication. H. Stager. (Schweizer Archiv, 1949, vol. 15, Apr., pp. 97–116). The present state of knowledge of friction and lubrication is reviewed, beginning with lubrication conditions in plain bearings with molecular-physical boundary-layer formation, continuing with high-pressure lubrication, and concluding with the machining and plastic deformation of metals. The author's work with an apparatus for measuring the viscosity of lubricants in tension and in shear is reported. It is shown that the adhesive strength of certain water-in-oil emulsions is so great that metal particles are torn out of the metal surfaces when determining the tensile viscosity of the emulsion.—R. A. R.

The Measurement of Kinetic Boundary Friction, or the Experimental Investigation of "Oiliness." J. R. Bristow. (Institution of Mechanical Engineers, Apr., 1949, Advance Copy, pp. 8–16). An instrument has been designed which allows the variation of boundary friction with velocity to be determined at very low velocities. Results are presented illustrating the influence of temperature, material nature of the surfaces, and surface finish on kinetic boundary friction for a variety of lubricants. These allow the boundary lubricating quality, or 'oiliness' of lubricants to be assessed in some measure. It is suggested that curves of the coefficient of boundary friction *versus* velocity constitute the correct form in which the results of measurements of kinetic boundary friction should be stated.—R. A. R.

The Surface Roughness of Bearing Surfaces and Its Relation to Oil Film Thickness at Breakdown. A. Cameron. (Institution of Mechanical Engineers, May, 1949, Advance Copy). The relation of surface roughness to allowable film thickness was studied quantitatively with a simple Michell pad apparatus. The pads were faced with white metal and ran against mild steel collars. There was little difference in the behaviour of any of the lubricants (water, soap solution, paraffin, and light oil), except that the aqueous lubricants would not run with very finely finished steel surfaces. The onset of metal to metal contact was detected by an increase in the frictional drag, and by the change in electrical conductivity between pad and collar. There is, for this system at any rate, a quantitative relation between the total surface roughness of the rubbing surfaces and the calculated oil film thickness both at the initial metal to metal contact and seizure. Initial contact occurs when the outlet film thickness, calculated from normal hydrodynamic theory, falls to three times the maximum surface roughness and seizure occurs when it is double the average roughness.—R. A. R.

Oxidation of Lubricating Oils at High Temperatures. G. H. Denison and O. L. Harle. (Industrial and Engineering Chemistry, 1949, vol. 41, May, pp. 934–937). Techniques have been developed for studying the rates at which lubricating oils take up oxygen at temperatures up to 340° C. The rates of oxygen absorption in the 170–340° C. range obey the Arrhenius law within the accuracy of the experimental method. The overall rate of absorption at 290° C. in the piston area of an engine was 200,000 times that at 90° C. in the sump.—R. A. R.

Antiseizure Properties of Boundary Lubricants. R. E. Thorpe and R. G. Larsen. (Industrial and Engineering Chemistry, 1949, vol. 41, May, pp. 938–943). The variation of the coefficient of friction with velocity was examined under boundary conditions in order to define the requirements of boundary lubricants and to establish the validity of Beeck's definition of boundary friction. The friction of effective boundary lubricants is independent of sliding velocity in the low speed range (0·1 to 8 cm./sec.), or decreases very slightly as the speed approaches zero. Poor boundary lubricants are characterized by relatively high friction, often increasing at low sliding velocities to values characteristic of unlubricated surfaces.—R. A. R.

Oxidation of Lubricating Oils. G. H. Denison and P. C. Condit. (Industrial and Engineering Chemistry, 1949, vol. 41, May, pp. 944–948). Various dialkyl selenides have been prepared and shown to be markedly more effective inhibitors of the oxidation of oil than the corresponding sulphides. The dialkyl selenides are far more active than the diaryl compounds.—R. A. R.

Complex Thiophosphoric Amides as Lubricant Additives. J. D. Bartleson and M. L. Sunday. (Industrial and Engineering Chemistry, 1949, vol. 41, May, pp. 948–951). Some reactions of phosphorus pentasulphide with a secondary aliphatic amine of high molecular weight and the reaction of these products with metal hydroxides are described. The structures of the reaction products are discussed. A typical product was investigated as an addition agent for lubricating

oils. It was very effective as a detergent, a depressor of the pour point, and a corrosion inhibitor, and it slightly improved the viscosity index.—R. A. R.

Hypoid Gear Lubricants. C. F. Prutton and P. A. Asseff. (Industrial and Engineering Chemistry, 1949, vol. 41, May, pp. 960–962). A brief report is presented on the development of lubricants for hypoid gears with data on the properties of some typical hypoid-gear lubricants.—R. A. R.

WELDING AND FLAME-CUTTING

Standard Welding Terms and their Definitions. (American Welding Society, May, 1949, Pamphlet). This publication contains more than 500 terms and 57 illustrations defining welding terms agreed by the American Welding Society. With it is issued a standard master chart of welding processes and process charts.—R. A. R.

Comparison of Methods for Determining Surface Area and Other Particle Size Data of Fine Powders, Particularly Welding Electrode Powders. P. D. Blake. (Journal of the Society of Chemical Industry, 1949, vol. 68, May, pp. 138–148). The methods of H. Heywood and of D. S. Roller, and a modification of the latter, for determining the surface area of fine powders are compared. The hydrometer method of size determination in the subsieve range is described and the data obtained by it from a study of standard welding electrode coating powders are used to compare the three methods. It is concluded that the modification of Roller's method gives the closest approximation to an absolute determination of the surface area for data derived from the hydrometer method of particle size determination.—R. A. R.

The Hot Electrode, a New Aid to Local Heat-Treatment and Heating. H. Schottky and F. H. Müller. (Schweissen und Schneiden, 1949, vol. 1, May, pp. 78–81). An electrode with a special coating has been developed as a means of applying intense local heat to a steel surface. Practically all the metal of the electrode is oxidized and the coating forms a semi-liquid slag on the surface being treated. In multi-run welds on thick plates the structure of the last run remains coarse so that the whole boiler or vessel has to be normalized. The deposition of a slag coating by the heating electrode on top of the last run not only causes the latter to solidify with a fine crystalline structure but also very greatly reduces the hardness of the heat-affected zone, and in some cases subsequent heat-treatment of the whole vessel is not necessary.

Investigation of Several Types of Electrodes for Electric Arc Welding of Type "E. Ya. I-T." Steel. E. M. Lapitskaya and I. N. Gerasimenko. (Avtogennoe Delo, 1948, Sept., pp. 12–14 [in Russian]: [Abstract] Metals Review, 1949, vol. 22, Feb., p. 48). Several methods are proposed to avoid the intercrystalline corrosion of nickel–chromium steel of the 18/8 and 25/20 types when ordinary electrodes are used.

The Part Played by Oxygen and Nitrogen in Arc Welding. J. D. Fast. (Philips Technical Review, 1948, vol. 10, July, pp. 26–34: [Abstract] Metals Review, 1949, vol. 22, Feb., p. 48). The action of oxygen and nitrogen on iron and steel is dealt with as an introduction to a discussion of the function of the coating of welding electrodes. The solubility of oxygen and nitrogen in liquid and solid iron is estimated and their harmful and beneficial effects in electric welding are noted.

The Absorption of Nitrogen when Arc-Welding with Bare Alloyed Electrodes in Comparison with Similarly Alloyed Coated Electrodes and with Various Commercial Electrodes. F. Bischof. (Die Technik, 1949, vol. 4, May, pp. 229–234). The effects of electrode composition and electrode coating composition on the nitrogen absorption of welds were investigated. Increasing manganese in the electrode increased the amount of nitrogen absorbed; chromium had the same effect, and with both alloying elements present the effect appeared to be additive. Nickel had the opposite effect. The addition of nickel to a chromium-bearing weld did not reduce the absorption of nitrogen. The addition of nickel to high manganese–chromium uncoated electrodes either very slightly reduced or did not affect the amount of nitrogen absorbed.—R. A. R.

Hydrogen in Arc Welding. J. D. Fast. (Lastechniek, 1949, vol. 15, May, pp. 220–224). [In Dutch].

Manual Submerged Arc Process Simplifies Welding for Many Uses. C. G. Herbruck. (Materials and Methods, 1949, vol. 29, June, pp. 64–66). A simple manually controlled arc welding process is described in which a bare welding wire $\frac{5}{64}$ in. in dia. is fed through a granular flux which is simultaneously supplied to the weld. Examples of its application are described and illustrated.—R. A. R.

The Laboratory in Resistance Welding. F. R. Hensel and E. F. Holt. (Welding Journal, 1949, vol. 28, June, pp. 531–540). The Resistance Welding Division of P. R. Mallory and Co. maintains a laboratory for the testing of the electrical and mechanical properties of the firm's production, and for the study of resistance welding problems in general. The laboratory and some of the work carried out are described. J. P. S.

MACHINING AND MACHINABILITY

Hot Spot Machining at Work Temperatures of 500°, 1000° and 1500° F. S. Tour and L. S. Fletcher. (Iron Age, 1949, vol. 164, July 21, pp. 78–89). Experiments have been carried out on the lathe turning of carbon and low-alloy steels and high-temperature-resisting nickel–chromium–cobalt material, surface-heated by induction or by gas to temperatures between 500° and 1500° F. (260° and 815° C.). Greatly increased feed and speed of cutting are possible, associated with reduced torque and thrust on the tool. Ratios of material removed hot to material removed cold are, for low-carbon steel, 3·3 : 1, for medium-carbon steel 3·9 : 1, and for low-alloy steel, 2·6 : 1. In the case of the heat-resisting alloy, whereas the recommended rate for cold machining is 2·1 cu. in./hr., by hot machining 270 cu. in./hr. could be removed.—J. P. S.

Cutting Temperature and Methods for Its Determination. A. P. Gulyaev. (Zavodskaya Laboratoriya, 1949, vol. 15, June, pp. 717–723). [In Russian]. Methods available for the determination of the temperature distribution in a cutting tool during its use are considered. An approximate method is described which is applicable to tool steels (carbon, alloy, and high-speed) and, to a lesser extent, to hard alloys. The method consisted of the following stages: (*a*) Thorough study of the red-hardness of the tool material; (*b*) the exposure of the tool to the action of heat by using it for cutting; and (*c*) the study of the residual effects of the heat generated on the hardness of the various parts of the tool. Some consequences of interest to the theory of cutting, which emerged from experiments carried out using the above method, are mentioned.—S. K.

Influence of Steel Hardness in Face Milling. J. B. Armitage and A. O. Schmidt. (Transactions of the American Society of Mechanical Engineers, 1949, vol. 71, July, pp. 413–419). An illustrated description is presented of face-milling tests on seven different kinds of steel having hardnesses of approximately 200, 300, and 400 Brinell. The steels used were NE 8744, NE 9445, SAE 1055, SAE 3140, SAE 4150, SAE 4150 sulphurized, and SAE 4335. At any given hardness, power requirements varied little, but the life of the cutting tool varied considerably. Machining at an elevated temperature was found to decrease the power required and to permit good cutter life. It is concluded that, in general, the harder a steel is, the greater will be the work required to machine it and the shorter will be the tool life if conditions of set-up, cutting speed, and feed are the same as those used for identical, but softer, steels. A reduction in cutting speed and use of finer feed will increase tool life when milling harder steel.—J. C. R.

Mechanics of Formation of the Discontinuous Chip in Metal Cutting. M. Field and M. E. Merchant. (Transactions of the American Society of Mechanical Engineers, 1949, vol. 71, July, pp. 421–430). A report is presented on the essentials of the geometry and mechanics of discontinuous chip formation.—J. C. R.

Constant-Pressure Lathe Test for Measuring the Machinability of Free-Cutting Steels. F. W. Boulger, H. L. Shaw, and H. E. Johnson. (Transactions of the American Society of Mechanical Engineers, 1949, vol. 71, July, pp. 431–446). This test evaluates materials on the basis of the feed resulting from a fixed horizontal tool pressure and requires only a short testing time. A description of the lathe and auxiliary equipment is given, and typical testing results are presented. The data indicate that the method has a high order of sensitivity and results obtained were in adequate agreement with large-scale commercial machining operations. The test appeared to offer a means of rating the cutting quality of free-cutting steels rapidly and reliably using small samples.

Machining Malleable Iron. M. Tilley, W. M. Albrecht, L. R. Spann, and J. H. Lansing. (Iron Age, 1949, vol. 163, June 2, pp. 72–76). The machining of a number of

automobile parts is described, details of cutting speeds, tools, and lubricants being given.—J. P. S.

Machining of Heat Resistant Steels. N. N. Zorev. (Stanki i Instrument, 1948, vol. 19, Sept., pp. 16–18 [in Russian]: [Abstract] Metals Review, 1949, vol. 22, Feb., p. 45). This is an account of an investigation on the machinability of a steel containing 0·52% carbon, 13·75% nickel, 13·22% chromium, 3·78% tungsten, 0·64% manganese, 0·4% silicon, 0·20% sulphur, and 0·016% phosphorus.

Machining Stainless Steel. L. F. Spencer. (Iron Age, 1949, vol. 164, July 7, pp. 83–89; July 28, pp. 64–68). The author first describes the various types of stainless steel, and the cutting rates that can be expected when machining them: he then passes on to consider the design of tools, including their side, top, and back rakes and clearances, and the provision of chip-breakers. He next describes the type of drill required for drilling, and in the second part of the article, studies the operations of threading and tapping. Finally, he describes the most suitable types of cutting oil.—J. P. S.

The Advantage of Cutting Tools Working in Compression. F. Eugène and R. Dufaud. (Comptes Rendus, 1949, vol. 229, July 4, pp. 53–54). Machine tools are usually so arranged that they work under flexural stresses. Recently, tool holders have been designed to allow the tools to work in compression, the axis of the tool being parallel to the principal stress of cutting. The benefits of the latter arrangement are enumerated.—A. E. C.

Electrolytic Tool Sharpening. S. E. Noskov. (Stanki i Instrument, 1948, vol. 19, Oct., pp. 20–22: [Abridged Translation] Engineers' Digest, 1949, vol. 10, Mar., pp. 95–97). Gussev's method for the electrolytic sharpening of tools is described; the same wheel is used, in this process, for all stages of sharpening and it can have a lower hardness than the tool to be sharpened. The tool and the grinding wheel are connected in a D.C. circuit, the wheel being the cathode and the tool the anode. Voltage and current values are given for roughing, coarse-grinding, and finishing. The electrical equipment which can be used for this process is discussed.—R. F. F.

CLEANING AND PICKLING

Polishing—Recommended Procedures for Ferrous and Non-Ferrous Metals. G. F. Weill. (Metal Industry, 1949, vol. 74, May 20, pp. 407–409; June 3, pp. 442–444). The importance of selecting the correct grade of abrasive for the first polishing operation is stressed and polishing sequences for cast iron, carbon steel, stainless steel, and non-ferrous metals are described.—R. A. R.

Polishing Stainless Steels. (Galvano, 1949, vol. 17, July, pp. 9–10). The technique of polishing stainless steels is described and two acid pickling solutions are given.—R. F. F.

An Electrolyte for the Electrolytic Polishing of Steel. W. Eilender, H. Arend, U. Eggers, and F. Sadrazil. (Metalloberfläche, 1949, vol. 4, Apr., Section A, pp. 88–90). The development of a new electrolyte for the electrolytic polishing of steel and of plated coatings, including chromium, is described. Tests on iron specimens with a surface of 10 sq. cm., with 3 amp. at 10–8·5 V. gave a polishing time of 10 min. The composition of the bath was 20% (by weight) H_2SO_4 (sp. gr. 1·84) and 80% H_3PO_4 (sp. gr. 1·70) with a few c.c. of water. The bath required changing when the iron content reached 12–15 g./l.—R. A. R.

Application of the Radioactive Tracer Technique to Metal Cleaning. J. C. Harris, R. E. Kamp, and W. H. Yanko. (A.S.T.M. Bulletin, 1949, May, pp. 49–52). For the determination of the efficiency of a metal-cleaning process which involves first, boiling in sodium metasilicate plus a commercial wetting agent, and then two rinses in boiling water, followed by one in running cold water, a compound containing the radioactive isotope of carbon, C^{14}, was added to the lubricating oil used as 'soil' in the experiment. This compound consisted of N, N-di-*n*-butyl stearamide, the C^{14} being substituted in the amide group. Whereas neither gravimetric nor fluorimetric estimation revealed any retained 'soil,' survey with the Geiger counter indicated the retention of some of the radioactive substance: repeated washing removed it. This was attributed to preferential adsorption of the radioactive substance; the substance has also the undesirable property of settling slightly from the lubricant, and a more suitable compound is being synthesized.—J. P. S.

Metal Surfaces: Their Preparation and Painting. G. Diehlman, A. J. Eickhoff, and T. G. Wills. (Official Digest of the Federal Paint and Varnish Producers Club, 1948, No. 280, pp. 357–367: British Abstracts, BII, 1949, Jan., col. 105). The painting of iron and steel, galvanized iron, and aluminium and its alloys is considered. Emphasis is laid upon the surface preparation of the metal before painting.

Flame Cleaning: A Modern Method of Preparing Steel for Painting. G. W. Rigby. (Yearbook of the Coke Oven Managers' Association, 1949, pp. 199–218). *See* Journ. I. and S.I., 1949, vol. 160, Oct., p. 228.

Preparation of Metal Surfaces Preparatory to Finishing. V. M. Darsey. (Paint and Varnish Production Manager, 1949, vol. 29, No. 3, pp. 63–70: ZDA Abstracts, 1949, vol. 7, May, p. 67). This is a survey of general metal cleaning and finishing techniques and notes on the preparation of various metals and their alloys.

Special Steels and Their Preparation for Enameling. F. R. Porter. (Better Enameling, 1949, vol. 20, Jan., pp. 17–19).

Some Comments on the Cleaning of Sheet-Metal Pressings Prior to Enamelling. P. G. Patten. (Sheet Metal Industries, 1949, vol. 26, June, pp. 1297–1301). Recommendations on the use of organic solvents, the application of alkali cleaners in tanks and by spraying, and the evaluation of alkali cleaners are made.—R. A. R.

Chemical Pickling of Iron and Its Alloys. J. Liger. (Galvano, 1949, vol. 17, Aug., pp. 9–11). The various alkaline and acid pickle liquors which can be used for pickling iron and its alloys, are discussed.—R. F. F.

Sodium Hydride Pickling. A. Normand. (Galvano, 1948, vol. 17, Apr., pp. 9–10). A low-temperature pickling process, using sodium hydride, which has been developed in America by the Dupont de Nemours Company, is described. Iron and steel, precious metals, chromium, nickel, and copper can be pickled with sodium hydride.—R. F. F.

Hot-Dip Galvanising. T. B. Crow. (Metallurgia, 1949, vol. 39, Apr., pp. 298–302; vol. 40, May, pp. 14–16). Investigations to develop a practical technique for the hydrochloric acid pickling of steel window frames before galvanizing are described. The procedure for pickle control is summarized as follows: (1) Maintain acid strength for as long as possible, subject to a voluntary cessation of acid additions after the iron concentration has reached 100 oz./cu. ft. (2) Make good the drag-out by 'topping up' with concentrated acid. (3) When the acid concentration is very low (say 10–20 oz./cu. ft., and the iron is over 100 oz./cu. ft.) the bath may be operated 'to the limit' until the iron reaches 120 oz./cu. ft. (4) An exhausted bath can always be used for stripping. Convenient forms of bath depth gauge and bath sampling apparatus are described in appendices.—R. A. R.

Eliminating Pickling Room Fog. J. T. Bolger. (Iron Age, 1949, vol. 163, May 26, pp. 79–81). To avoid the occurrence of fog caused by the condensation of vapours from pickling tanks, a blanket of hot air is projected across the top of the tanks.—J. P. S.

PROTECTIVE COATINGS

Principles of Electroplating. H. E. Linsley. (Machinist, 1949, vol. 93, July 9, pp. 383–398). The general electroplating methods are discussed under three main headings: Cleaning procedure, plating procedure, and plating equipment.—R. F. F.

Tank Linings and Insulating Materials. V. Evans. (Journal of the Electrodepositors' Technical Society, 1949, vol. 24, pp. 129–143). The paper reviews the principal materials used in the construction of tank linings to deal with corrosive conditions such as those found in pickling and plating operations. Insulating methods for rack and jig coatings are also considered. Emphasis is placed on more recent and less conventional methods of tank lining, including the use of some plastic materials. Consideration is also given to suitable combinations of corrosion-resistant materials to provide complete structures of good mechanical strength and chemical resistance.—R. A. R.

Development of Pipe Line Coatings and Mechanical Application Methods. S. D. Day. (Corrosion, 1949, vol. 5, July, pp. 221–226). Materials and methods used in the coating of gas and oil pipelines for sub-surface use are described.—J. P. S.

Cost and Benefits of Conservation of Cast Iron and Steel Products by Control of Corrosion. J. E. Carrière and C. A. Lobry de Bruin. (United Nations Scientific Conference on the Conservation and Utilization of Resources, Lake Success,

1949, Aug. 17–Sept. 6 ; Advance Copy). The methods used for protecting gas and water pipes in Holland, where, because of the low-lying nature of the country, fresh and salt water occur very near the surface, are described. These methods rely mainly on the application of blown asphaltic bitumen, in single coatings where conditions are least severe, and ranging up to eight coatings, including impregnated fibre cloth, where conditions are at their worst. The economic savings, under Dutch conditions, are analysed. Methods of soil testing are also described.—J. P. S.

Centrifugal Lining of Pipe. R. C. Mahon. (Mining Engineering, 1949, vol. 1, May, Section 1, p. 19). A brief account is given of a method for the centrifugal lining of piping with concrete, the lining being $\frac{1}{4}$ in. thick up to a pipe diameter of 8 in. and $\frac{3}{8}$ in. thick for larger pipes. The lined piping is suitable for the discharge of acid mine waters.—J. C. R.

The Production and Uses of Clad Steels. F. R. Pattison. (Journal of the Birmingham Metallurgical Society, 1949, vol. 29, June, pp. 85–110). The manufacture of clad steels by casting, inter-melting, fusion-, arc-, and resistance welding, and bonding by heat and rolling pressure is briefly described, and their selection, handling, and fabrication are discussed.

R. A. R.

Bonding of Steel to Cast Aluminium and Uses Thereof. C. E. Stevens, jun. (Metal Progress, 1949, vol. 55, Mar., pp. 326–331). The technique of bonding steel to cast aluminium and the properties of the joint are discussed. A steel cylinder liner (of an aeroplane engine) chemically bonded to aluminium exterior fins had vastly better heat transfer than the conventional finned steel (integral) construction. The process, known as ' Al-Fin,' has been extended successfully to the manufacture of brake drum and power-tube radiators, to aluminium gearing with steel hubs, and to steel-backed bearings.—J. C. R.

The Protection of Metals against Corrosion. T. P. Hoar. (Journal of the Royal Institute of British Architects, 1949, vol. 56, Apr., pp. 276–281). The mechanism of the corrosion of metals used outside and inside buildings is explained and the advantages and characteristics of different paint and metal coatings are pointed out. A new electroplating that is becoming established is white bronze, a 60/40 copper–tin alloy, white in colour, and more like silver than chromium.

Coloring of Metals. N. B. Bagger. (Materials and Methods, 1949, vol. 29, June, pp. 67–82). Comprehensive tables are described giving particulars of processes of colouring iron and steel and some non-ferrous metals.—R. A. R.

Treatment of Plating Mill Wastes. M. U. Priester. (Iron Age, 1949, vol. 164, July 14, pp. 105–106). The treatment of wastes from an electroplating establishment is described. Plating and cyanide-bearing solutions and rinses are treated with calcium hypochlorite and sulphuric acid and settled before decantation of the clear solution and lagooning of the sludge : wastes from tumbling barrels and cleaning and degreasing operations receive a treatment to free them from oil before being mixed with the other solutions to receive the same treatment.—J. P. S.

Plating Waste Disposal. J. L. Bleiweis. (Iron Age, 1949, vol. 163, June 16, pp. 78–83). Plating wastes may contain hexavalent chromium, having an economic value, acids, alkalis, cyanides, and materials in suspension. Methods are given for recovering chromium in various forms, and for reducing the harmful contents of the wastes to a point at which they can be discharged into waterways.—J. P. S.

Hard Chrome Plating Milling Machine Quills. W. Schmidt. (Iron Age, 1949, vol. 163, June 16, pp. 96–97). These parts of milling machines run up and down, without revolving, in ground and lapped cast-iron housings ; the manufacturers find hard chromium plating gives excellent wear-resisting properties, negligible wear in 14 years being reported in one instance. The plating procedure is described and the cost of producing a plated quill is compared with that of producing one carburized, hardened, and ground.—J. P. S.

Wearing Conditions and Quality Control of Hard Chromium Plating. P. Morisset. (Revue Générale de Mécanique, 1949, vol. 33, Apr., pp. 142–146). The various conditions affecting the wear-resisting properties of hard chromium coatings are discussed ; these include friction, lubrication, and temperature. The relation between the hardness and the wear-resistance of hard chromium is studied and the hardness tests suitable for hard chromium coatings are described.—R. F. F.

Hard Chromium Plating with Superimposed High-Frequency Alternating Current. F. Muller, W. Eilender, and K. M. Wagner. (Archiv für Metallkunde, 1949, vol. 3, Apr., pp. 135–140). The validity of the claims of H. Hausner respecting his process of chromium plating with a superimposed alternating current were investigated. Under the plating conditions described the results indicated that there was no improvement in current efficiency, striking power, or quality of the plating.—R. A. R.

Impurities in the Nickel Baths. J. Salauze. (Galvano, 1948, vol. 17, Sept., pp. 9–17). The types of impurities and their effect on the nickel bath, methods of preventing contamination of the bath, and of removing impurities are discussed.—R. F. F.

Continuous Galvanizing of Strip Steel. H. W. Lynn. (Iron Age, 1949, vol. 164, July 21, pp. 96–100). A new continuous hot-dip galvanizing line installed at Weirton Steel Co. operates at 300 ft./min., and deals with material up to 48 in. wide. After cold rolling, the strip is electrolytically cleaned, re-coiled, annealed in coil, rolled flat without hardening, and re-coiled. Coils then pass to the galvanizing line ; the strip goes through a pinch roll and the end is sheared and welded to the tail of the preceding coil ; next comes a side shear, and a looping pit. The strip is then led through water scrubbers, an acid dip, and a rinse-spray, and passes on through a bath of aqueous flux to a radiant-tube furnace which dries the flux and preheats the strip. It then passes through the zinc kettle, over a cooling tower to the cooling table. After a further cleaning operation it passes through a roller leveller and is sheared to length. It is now ready for inspection, dispatch, or corrugating.—J. P. S.

Hot-Dip Galvanising.—The Possibilities of Gas Heating. (Gas Heat in Industry, 1949, May, pp. 231–237). Methods of heating hot-dip zinc galvanizing baths are compared and the use of gas is discussed. A method of recent design is described in which immersion gas heaters are fixed in pockets containing molten lead along the sides and ends of the bath, the lead transmitting the heat to the zinc through the bath walls. The gas consumption for a typical week in which $11\frac{1}{2}$ tons were galvanized in 117 working hours was 90,000 cu. ft./ton (at 530 B.Th.U. cu. ft.). The temperature of the lead pockets was closely controlled at 465° C.—R. A. R.

The Effect of Titanium in Iron when Hot-Dip Galvanizing. H. Bablik, F. Götzl, and R. Kukaczka. (Zeitschrift für Metallkunde, 1949, vol. 40, May, pp. 176–179). Although closely related in the Periodic System, the elements silicon and titanium in iron have very different effects on the iron–zinc reaction in hot-dip galvanizing. About 0·7% of silicon increases tenfold the amount of iron taking part in the iron–zinc reaction as compared with silicon-free iron. There is no increase in the reaction velocity when titanium is added, and this element causes larger and more regular zinc spangles to form.—R. A. R.

The Importance of the Nitrogen in the Iron in Hot-Dip Galvanizing. H. Bablik, F. Götzl, and R. Kukaczka. (Zeitschrift für Metallkunde, 1949, vol. 40, Apr., pp. 141–147). The iron–nitrogen constitutional diagram up to 8% nitrogen and the effect of a nitrided iron surface on subsequent hot-dip galvanizing were studied. When nitriding at below 590° C. the structure consists mainly of acicular nitrides whereas, at above 590° C., it consists of either grain-boundary cementite or the eutectoid braunite, similar to pearlite, as an external layer or as islands below the surface. The nitriding makes the iron more soluble in acid and loosens the surface, probably giving better adherence of the zinc coating. The velocity of the iron–zinc reaction is reduced by nitriding.

R. A. R.

Silver-Plating of Steel. J. Liger. (Galvano, 1949, vol. 18, Feb., pp. 9–11). The technique and the baths used for silver plating steels are discussed.—R. F. F.

New Torch Licks Problem of Flame-Spraying Polythene. W. B. De Long and E. V. Peterson. (Chemical Engineering, 1949, vol. 56, June, pp. 123–125). A gun is described in which polymerized ethylene in powder form is sprayed by compressed air through a nozzle on to a preheated metal surface.—R. A. R.

Some Recent Developments in Synthetic Resins for Protective Coatings. L. R. Whiting. (American Paint Journal, 1948, vol. 32, No. 30, pp. 72–82 : British Abstracts, 1949, BII, Jan., col. 105). The importance of the surface preparation of steel before painting is stressed. A wash primer,

consisting of a polyvinyl butyral resin lightly pigmented with zinc chrome with H_3PO_4 added, gives good results. Best results are obtained by using a special resin system over the wash primer.

How to Specify Organic Finishes. A. L. Alexander. (Product Engineering, 1949, vol. 20, July, pp. 141–145). Organic coatings are classified into varnishes, paints, enamels, lacquers, shellac, bituminous coatings, and rubber and rubber-like coatings. Their characteristic properties and applications, and six typical industrial finishing formulations are given.—R. A. R.

Building Organic Protective Coatings to Special Requirements. K. G. Compton. (Corrosion, 1949, vol. 5, May, pp. 148–150). Examples are given of the development of protective coatings for special purposes such as (*a*) black enamel for telephone parts, (*b*) a maintenance paint for plating shops, and (*c*) a coating for steel and aluminium storage bins for preserving ordnance equipment in tropical and other climates.

Latest Developments in Phosphate Coating Methods and Technique. H. A. Holden. (Journal of the Electrodepositors' Technical Society, 1949, vol. 24, pp. 111–128). Recent developments in phosphate coating methods are surveyed. An account is included of the Parco-Lubrizing process using iron/manganese phosphating solutions, and the use of phosphate coatings to assist lubrication of bearing surfaces and reduce the wear of cylinder liners, pistons, piston rings, camshafts, worm gears, and similar parts.—J. C. R.

Sprayed Metals. A. E. Williams. (Power and Works Engineering, 1949, vol. 44, Apr., pp. 111–113). The metal-spraying process and some of its applications are described. R. A. R.

Metallizing in Relation to Marine Engineering. J. B. Stiles. (Welding Journal, 1949, vol. 28, June, pp. 541–558). The author first describes the wire type of metal-spraying equipment, and methods of surface preparation, such as sand-blasting, rough-turning, Fuse-Bonding, Sprabonding, spray-welding, etc., and passes on to finishing of sprayed surfaces by turning and grinding. He then gives a number of cases of general application of the process in the repair of worn parts, the rectification of machining errors and the sealing up of casting porosity; examples from the marine field are the repair of tailshafts, pump shafts, and stern tubes. The value of sprayed zinc coatings in preventing sea-water corrosion is then established as due to their anodic sacrificial effect, with the corollary that coatings cathodic to iron and steel, such as copper or tin, must be rendered impervious to the corrosive medium, either by the application of heavy coatings or by impregnation with sealing compounds. Finally, the considerations involved in zinc-plating a ship's hull are examined.—J. P. S.

Metal Spraying in the Maintenance of Mine Equipment. R. Mansell. (Canadian Mining Journal, 1949, Apr., pp. 75–78). Examples of successful repairs to mining machinery parts by spraying metal on to the prepared worn areas are described and illustrated.—R. A. R.

Designing with Metallized Structures. J. E. Wakefield. (Product Engineering, 1949, vol. 20, May, pp. 117–121). The preparation of parts for metallizing, the properties of the coating, and the advantages of the process in many applications are discussed.—R. A. R.

The Leaching Rate and Behaviour of Antifouling Compositions. H. Barnes. (Journal of the Oil and Colour Chemists' Association, 1948, vol. 31, pp. 455–470: Chemical Abstracts, 1949, vol. 43, May 10, col. 3631). The most efficient composition is one where the poison is released at the required critical leaching rate for overall prevention of fouling. Poison lost at a higher rate will be wasted; at a lesser rate, the composition will foul. Once the critical leaching rate is attained the maximum life at a given weight per unit area is fixed; this may be calculated by dividing the poison content in γ per sq. cm. by the critical leaching rate. Increased efficiency results from the use of more effective poisons.

A Control System for Spray Painting. E. A. Zahn. (Iron Age, 1949, vol. 163, May 19, pp. 96–99). Five points to be observed in the attainment of improved finish and reduced finishing cost in spray painting are: (1) Paint utilization should be studied to determine whether dip or flow-coat methods can be used instead of spraying. (2) The job should be put on a piecework basis. (3) All spray equipment should be standardized. (4) Control of the paint should be instituted to see that solvents are not wasted. (5) Operators should be trained in efficient spray motion.—J. P. S.

Protective Coatings for Exposed Iron Structures. H. G. Wolf. (Farbe und Lack, 1948, vol. 54, pp. 124–125: British Abstracts, BII, 1949, Jan., col. 105). Various types of protective coatings are discussed with regard to their anti-corrosive and anti-fouling properties, chemical inertness, light-fastness, and heat resistance. Unsaponifiable synthetic products, such as chlorinated rubber, polyvinyl chloride, and benzylcellulose are preferred as media. The primer should contain the same medium as the top coat, and also the same pigments or iron oxide.

Protection of Steels by Means of Water-Excluding Coatings. J. Göbel. (Verfkroniek, 1948, vol. 21, pp. 127–130: British Abstracts, BII, 1949, Jan., col. 106). Methods of overcoming defects such as cracking or excessive flow, associated with the weathering of bituminous protective coatings, by the use of hot-applied blown bitumens, bitumen emulsions, or bitumen solutions containing fillers, *e.g.*, fibrous asbestos, are discussed. Bitumen paints containing aluminium powder offer the most promise.

Mechanism of Pigmentation of Anti-Corrosive Paints and Varnishes. H. Wagner. (Paint and Varnish Production Manager, 1949, vol. 29, No. 3, pp. 63–70: ZDA Abstracts, 1949, vol. 7, May, p. 66). This is a review of published information on the anti-rusting properties of pigments.

POWDER METALLURGY

Present-Day Processes for Producing Iron Powder, their Raw-Material Basis and Economy. H. Buchholtz. (Stahl und Eisen, 1949, vol. 69, Apr. 14, pp. 247–256). The technical and economic aspects of several modern processes for producing iron powder are critically examined. In Germany the greater part of the total production is by the Hametag eddy-mill process using wire and plate cuttings as raw materials; in other countries the processes are usually based on the reduction of ores and mill-scale. The importance of taking geographical aspects and the raw material and electricity supply situation into account when deciding what process to use is stressed.—R. A. R.

Recent Developments in Powder Metallurgy. Part 2—Products. P. Schwarzkopf. (Powder Metallurgy Bulletin, 1949, vol. 4, May, pp. 64–111). The review of the literature is continued in this Part which covers the products of powder metallurgy (*see* Journ. I. and S.I., 1949, vol. 163, Sept., p. 105). There are 300 references.—R. A. R.

A Compressibility Test for Metal Powders. F. V. Lenel. (A.S.T.M. Bulletin, 1949, May, pp. 52–56). Experimental data are reported on the influence of various lubricants upon the compressibility of metal powders. Copper and iron powders were used in these investigations. It was not found possible to base a standard compressibility test method on the results obtained.—J. C. R.

Electrical Resistivity Measurements on Iron–Silicon Compacts Prepared by the Powder Metallurgy Procedure. F. W. Glaser. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2660: Transactions of the American Institute of Mining and Metallurgical Engineers, 1949, vol. 185, pp. 475–480: Journal of Metals, 1949, vol. 1, Aug., Section 3, pp. 475–480). The author has studied, by electrical resistivity measurements, the progress of diffusion occurring in iron–silicon alloys. Iron, silicon, and iron–silicon powders of various mesh size and compositions were used and the samples were prepared by pressing and sintering. Complete diffusion was assumed to have been obtained when the electrical resistivity reached a constant value characteristic for that particular alloy. Electrical resistivity measurements were used to follow the progress of alloying under various conditions. It was found that, in the higher density ranges, electrical resistivity was a function only of the diffusion process and was greatly influenced by changes of the silicon concentration in a non-homogeneous iron–silicon alloy. Values for the specific electrical resistivity of iron–silicon alloys containing 6–50% silicon were determined. The resistivity curve shows a break between approximately 11 and 25% silicon. No major effects upon the electrical resistivity of an atmosphere containing nitrogen were detected for iron–silicon alloys.

Scope and Limitations of the Powder Metallurgy Process. E. R. Engstrand, K. M. Gleszer, and J. W. Polonetz. (Product

Engineering, 1949, vol. 20, July, pp. 123–127). Recommendations are made on the powder-metallurgy process regarding the limits of dimensions; design considerations such as proportions, projections, tolerances, wall thickness, draft, chambers, ribs, losses, screwthreads, holes, inserts, fins, and lettering; and subsequent operations, such as machining, coining, and hardening.—R. A. R.

The Design and Operation of Small Mills for Mixing Powder. D. Summers-Smith. (Metallurgia, 1949, vol. 39, Apr., pp. 309–311). Detailed descriptions are given of the design of two ball mills of 75 and 10 c.c. capacity, respectively, for mixing powders for preparing alloy samples. The larger mill has six separate milling chambers; it is driven at 60 r.p.m. and satisfactory mixing is obtained in about 12 hr.—R. A. R.

Sintered Hard-Metals as Wear-Resisting Materials. R. Kieffer and F. Benesovsky. (Werkstatt und Betrieb, 1949, vol. 82, May, pp. 145–148). Data are presented on the hardness and bend strength of a number of tungsten-carbide/cobalt and tungsten-carbide/titanium-carbide/cobalt sintered hard metals, and their applications are discussed.

R. A. R.

PROPERTIES AND TESTS

Metallurgical Inspection of Castings. R. R. Senz. (Foundry, 1949, vol. 77, June, pp. 82–83, 246–248). A brief survey is presented of destructive and non-destructive methods of inspecting castings.—J. C. R.

The Non-Destructive Testing of Materials. R. Lindemann. (Die Technik, 1949, vol. 4, Apr., pp. 151–160). Testing methods in use at the Materialprüfungsamt in Berlin since 1945 are reviewed, mention being made of tests with improvised apparatus.—R. A. R.

New Machines for Testing the Mechanical Properties of Materials. Kh. N. Dementev. (Zavodskaya Laboratoriya, 1949, vol. 15, June, pp. 733–737). [In Russian]. A description is given of a new machine for tensile testing under loads up to 5000 kg. Its accuracy, estimated largely by geometrical considerations, is favourably compared with that of some other testing machines. Some shortcomings of standard testing procedure in the U.S.S.R. are mentioned, these comments being supported and amplified in an accompanying editorial note, which also invites opinions on this topic.—S. K.

Effect of Speed of Testing on the Tensile Properties of Austenitic Stainless Steel Sheets. R. H. Heyer. (A.S.T.M. Bulletin, 1949, May, pp. 57–62). Experimental data derived from investigations in America in 1943, 1944, and 1948 on the effect of speed of testing on the tensile properties of austenitic stainless steel sheets, are presented.—J. C. R.

Measurement of the Number of Revolutions of Centrifugal Velocity Impact Machines, and of the Work Performed in the Fracture of the Specimen. P. G. Korolev. (Zavodskaya Laboratoriya, 1949, vol. 15, June, pp. 723–729). [In Russian]. In the testing machine considered, the specimen is fractured by the impact of a knife fixed to the rim of a disc rotating freely at a known speed. The loss of energy due to the impact is calculated from the corresponding decrease in speed, and the requirements of a device for automatically recording such decreases for the high frequencies encountered are discussed. A description is given of a 'beat-frequency' apparatus developed for this purpose by means of which the frequency of revolution just before the impact, as well as the change in frequency on impact, can be automatically recorded.—S. K.

A Metallographic Description of Fracture in Impact Specimens of a Structural Steel. M. Baeyertz, W. F. Craig, jun., and E. S. Bumps. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2635: Transactions of the American Institute of Mining and Metallurgical Engineers, 1949, vol. 185, pp. 481–490: Journal of Metals, 1949, vol. 1, Aug., Section 3, pp. 481–490). An account, illustrated by numerous photomicrographs, is presented of the nature of the fracture of specimens from a single structural steel shape, when subjected to impact tests at various temperatures. Fracture was found to be initiated by cleavage of ferrite grains situated in the general region of maximum triaxiality of stressing during the test. At the low-temperature end of the transition, ferrite cleavage occurred during or soon after the end of plastic deformation. It was often accompanied or preceded by shock twinning of the ferrite. At higher testing temperatures within the transition, shock twinning gave way to deformation by slip. In the high-temperature end of the transition, cold work apparently occurred both before and after the ferrite cleaved. A second stage of the fracture consisted of opening up the ferrite cleavages, deformation, and rupture of the metal between the cleavages. Experimental observations suggested that hard phases such as cementite and inclusions were not to blame for initiating ferrite cleavage in mild steel. There was evidence that pearlite could arrest ferrite cleavage and confine it to a ferrite band, at least in the middle and high-temperature end of the transition. The tendency for the initial ferrite cleavage to extend quite generally throughout the ferrite band and the directional extent of those bands in rolled mild steel products is suggested as a possible explanation of the difference in the energy values obtained with longitudinal and transverse specimens.

Discontinuous Crack Propagation. L. D. Jaffe, E. L. Reed, and H. C. Mann. (American Institute of Mining and Metallurgical Engineers, Technical Note No. 16: Transactions of the American Institute of Mining and Metallurgical Engineers, 1949, vol. 185, p. 526: Journal of Metals, 1949, vol. 1, Aug., Section 3, p. 526). A preliminary account is given of an investigation into the origin and propagation of cracks. An alloy steel (C 0·28%, Mn 0·74%, Si 0·20%, Cr 1%, Mo 0·49%, V 0·12%) was quenched from 1675° F. and tempered at 1150° F., and fractured by reversed bending. The fracture was examined microscopically and a suggested explanation of the propagation of cracks is offered.—J. C. R.

Mechanism of Flow for Solid Metals. H. Eyring, J. W. Frederickson, and D. McLachlan, jun. (Proceedings of the National Academy of Sciences, Washington, 1948, vol. 34, pp. 295–304: British Abstracts, BI, 1949, Jan., cols. 49–50). A general account of the structure of solids and mechanism of flow is given together with a classification of the imperfections of solid materials into general types. A detailed consideration is made of the imperfections created by fluctuations in d. The expression $(d\Delta d^2)/ = \Delta F^*/kT =$ constant is derived (ΔF^* = free energy of activation). From the constancy of $\Delta F^*/T$ for brass and SAE 1020 steel, it appears that neither the nature of the substance nor a difference in mechanism affects its value. Hence, it is concluded that the process of activation is one of crowding together of some molecules so as to create the holes or imperfections necessary for the flow process. The volume of the hole, Vh, into which the flowing unit jumps, appears to be temperature-dependent and Vh/T is approximately constant. Since calculation of Vh gives unreasonably low results, it is concluded that the material being compressed or rarefied is, for metals, the system of positive nuclei, *i.e.*, the atoms minus their conductance electrons. The nuclear compressibility is $\sim$ 1000 times that of the bulk metal. Metals with a large Vh will flow more easily than those with a low Vh. Preliminary investigation shows that $\Delta F^*/T$ is not independent of temperature and substance for non-metals.

The Plastic Deformation of Crystals According to Taylor's Theory. N. A. Brunt. (Metalen, 1949, vol. 3, Apr., pp. 169–175). [In Dutch]. G. I. Taylor's theory of dislocations is explained and the possibility of using it to explain the work hardening of metals is discussed. Finally, J. M. Burgers' theory is considered.—R. A. R.

A Reconsideration of Deformation Theories of Plasticity. D. C. Drucker. (Transactions of the American Society for Mechanical Engineers, 1949, vol. 71, July, pp. 587–592). Deformation theories of plasticity for strain-hardening materials are defined as those which postulate that the components of elastic and permanent strain are completely determined by the existing components of stress. The assumption is made that there is some criterion, in terms of stress, for loading or increase in plastic deformation. A common supposition of this nature is that the permanent strain will increase when the octahedral shearing stress increases. It is shown that no matter what the loading criterion, these deformation theories of plasticity lead to the unacceptable conclusion that large (finite) changes in the components of permanent strain may accompany infinitesimal increases in loading despite strain-hardening. Simple illustrations are given and it is demonstrated that an incremental, or so-called 'flow theory' eliminates this undesirable feature.

Fracture of Ductile Metals. J. E. Dorn. (Iron Age, 1949, vol. 164, July 7, pp. 90–95, 100). A number of theories on the plastic deformation and fracture of metals, and the degree

to which they account for the observed phenomena, are discussed. Much of the experimental data are not consistent with a fracture stress law but can be reconciled with a crystal damage hypothesis. The role of micro-cracks is significant in this connection, especially where combined stresses are involved, and the author considers that the rate of growth of these, and the possibility of their 'healing' under high hydrostatic pressures, should be investigated.—J. P. S.

On the Question of the Determination of Plasticity by the Method of Rolling into a Wedge. Iu. M. Chizhikov. (Zavodskaya Laboratoriya, 1949, vol. 15, June, pp. 737–739). [In Russian]. The author discounts the criticisms and suggested improvements to his method of determining plasticity made recently by I. K. Antsiferov. The method consists essentially of rolling a specimen of constant section into the shape of a wedge and determining the degree of reduction at which surface cracks appear; in this paper questions of specimen dimensions and method of preparation are dealt with. Illustrations of points made are provided by photographs of the macrostructure of ingots of different dimensions of corrosion-resisting steel and of a chromium–aluminium alloy, and by a diagram showing the variation with temperature of the limit of plasticity of a chromium–nickel alloy.—S. K.

Effect of Several Notches on Stressing—Stress-Relieving and Stress-Increasing Notches. A. Thum and O. Svenson. (Schweizer Archiv, 1949, vol. 15, June, pp. 161–174). The effects of various combinations of notches on the stress distribution in shafts and bars in tension and bending have been investigated using a small inductive static extensometer developed by the Materialprüfungsanstalt Darmstadt. Additional notches in a notched section, transverse to the direction of stress may (*a*) increase, (*b*) leave unchanged, or (*c*) alleviate the existing stresses, according to their shape and position. If the additional notches are in the zone already affected by the existing notches, the stress concentration will be considerably increased. If the additional notches are outside this zone and the total reduction in cross-section is not reduced by more than 20–30%, the stress peaks will remain practically unaltered. If, owing to their favourable shape and position, the additional notches reduce the shape factor more than the cross-section, the stress peaks will be slightly relieved.

More complicated cases, such as (1) a hole joining two notches, (2) two holes intersecting at right-angles, and (3) a notch at the point of application of a force, are considered, and a method of calculating the stress distribution is presented.—R. A. R.

A Report of Some Recent Research into the Measurement and Relaxation of Residual Stresses. W. Soete. (International Welding Conference, Brussels, 1948: Sheet Metal Industries, 1949, vol. 26, June, pp. 1269–1281: Welding Journal, 1949, vol. 28, Aug., pp. 354-s–364-s). This is an English translation of a paper which appeared in Revue de la Soudure, 1948, vol. 4, No. 3, pp. 159–169 (*see* Journ. I. and S.I., 1949, vol. 161, Feb., p. 156).—R. A. R.

The Revelation of Internal Stresses and Dislocations in Metals by the Photo-Elastic Behaviour of Silver Chloride. J. F. Nye. (Société Française de Métallurgie, Oct. 19, 1948: Revue de Métallurgie, Mémoires, 1949, vol. 46, June, pp. 371–375). Silver chloride is both transparent and very ductile; it can be drawn and folded to a considerable extent without breaking. For the investigation by photo-elastic methods of the internal stresses set up in metals when they are deformed plastically, the silver chloride is prepared in the form of rolled sheets 0·25 to 1 mm. thick; the specimens cut from them are first submitted to a crystallization treatment; then the resulting polycrystalline aggregate, free from stresses, is deformed plastically by simple drawing, folding, twisting, etc. Examples of the sort of effects obtained, as seen under the microscope by polarized light, are reproduced and discussed.—A. E. C.

Mohr's Circle and Its Applications to the Common Materials. B. Hedde D'Entremont. (Revue Générale de Mécanique, 1949, vol. 33, July, pp. 289–297). The theory of stresses and strains in materials is discussed, and the method, known as 'Mohr's circle,' used for the determination of normal stresses and maximum shear stress is described together with its applications.—R. F. F.

Mining Gear Failures from a Metallurgical Aspect. R. Jeffrey. (Transactions of the Institution of Mining Engineers, 1949, vol. 108, May, pp. 323–347). The results of metallurgical examinations of a large number of mining gear failures are examined. Design and fabrication accounted for 44% of the failures, poor materials for 25%, effects of service for 21%, and heat-treatment for 10%. Recommendations for reducing the number of failures are made.—R. A. R.

The Control of Surface Conditions by Fluorescence. M. Deribéré. (Revue Générale de Mécanique, 1949, vol. 33, June, pp. 251–256). After surveying the various methods which can be used for detecting surface defects, the author describes the fluorescent method. He gives details of the types of defects which can be detected by this method, of the materials which are required and of some practical results. The bibliography contains 13 references.—R. F. F.

The Statistical Nature of the Endurance Limit. J. T. Ransom and R. F. Mehl. (American Institute of Mining and Metallurgical Engineers, Technical Note No. 15: Transactions of the American Institute of Mining and Engineers, 1949, vol. 185, pp. 364–365: Journal of Metals, 1949, vol. 1, June, Section 3, pp. 364–365). Research work is being carried on to evaluate the statistical nature of endurance properties. Two methods have been adopted: (1) Straightforward statistical studies of cycles to fracture of a number of specimens at each of a series of stress levels; (2) a new abbreviated statistical method known as 'staircase testing,' which consists in testing a series of specimens, one by one, with the test stress for any given specimen automatically determined by whether the preceding specimen failed or ran out unbroken through ten million cycles. The stress for the first specimen is chosen at a level where 50% failures would be expected if a large number of specimens were tested, that is, at about one-half the tensile strength for steels. If the first specimen fails, the stress for the second specimen is decreased by one step, say 500 lb./sq. in. If the first specimen runs out unbroken the stress for the second specimen is raised by one step, and so on. It is claimed that fatigue properties of steels can be accurately appraised only by statistical methods.

J. C. R.

Influence of Type of Machine, Range of Speed, and Specimen Shape on Fatigue Test Data. P. K. Roos, D. C. Lemmon, and J. T. Ransom. (A.S.T.M. Bulletin, 1949, May, pp. 63–65). Fatigue tests were made on flat and round specimens of a SAE 4340 steel in pure reversed plane bending at 1800 r.p.m. on a Sonntag universal testing machine; and on round specimens acting as rotating beams on R. R. Moore fatigue machines at 1800 r.p.m. and 10,000 r.p.m. The data show that a higher endurance limit is obtained in pure reversed plane bending tests than in rotating beam tests; in round specimens subjected to pure reversed plane bending a higher endurance limit is obtained than in flat specimens similarly tested. In rotating beam tests performed at 10,000 r.p.m. a higher endurance limit is obtained than with tests performed at 1800 r.p.m.

Significant Stress and Failure in Static and Fatigue Loading. C. Lipson, G. C. Noll, and L. S. Clock. (Product Engineering, 1949, vol. 20, July, pp. 130–135). Formulæ are developed for calculating the effects of fillets, notches, and transverse holes on the fatigue strength of solid and hollow shafts.

A New High Speed Sheet Metal Fatigue Testing Machine for Unsymmetrical Bending Studies. G. R. Gohn and E. R. Morton. (American Society for Testing Materials, 1949, Preprint 31). A constant deflection fatigue testing machine capable of bend-testing 24 sheet metal specimens at a time, with 12 different values of mean stress, at testing speeds up to 3000 r.p.m. is described. Some results on phosphor bronze strip are presented.—R. A. R.

Fatigue Strength of Cast Crankshafts. H. R. Mills and R. J. Love. (Institution of Mechanical Engineers, Apr., 1949, Advance Copy). The paper gives the results of a systematic investigation into the effects of the proportions of a cast crankshaft on its bending fatigue strength.—R. A. R.

Fourteenth Progress Report of the Cooperative Investigation of Failures in Railroad Rails in Service and Their Prevention. R. E. Cramer. (University of Illinois, 1948, Engineering Experiment Station Bulletin, Reprint Series No. 39, pp. 3–8). The causes of rail failures classified as transverse fissures have been examined. Since controlled cooling was adopted in 1936 the number of this type of failure has been reduced to about 20 per year. Some of the failures were due to the heat of grinding, and others to shrinkage cracks in welds on built-up rail ends.—R. A. R.

Sixth Progress Report of the Shelly Rail Studies at the University of Illinois. R. E. Cramer. (University of Illinois, 1948, Engineering Experiment Station Bulletin, Reprint Series No. 39, pp. 9–18). A new design of testing machine which applies adjustable rolling loads to sections cut from rails is described and some results of tests are reported. The results indicate that heat-treated carbon steel rails and some alloy steel rails can be expected to give two or more times the service life of standard carbon steel rails.—R. A. R.

Third Progress Report on Corrugated Rails. R. E. Cramer. (University of Illinois, 1948, Engineering Experiment Station Bulletin, Reprint Series No. 39, pp. 19–20). The causes of a type of wear on rails known as 'corrugation' have been examined. No definite conclusions have been reached, but there is evidence that skids of locomotive driving wheels would bring the rail surface above the recrystallization temperature and rapid cooling by the mass of metal would result in hard spots which, with subsequent wear, become brighter than the remainder of the rail.—R. A. R.

Fatigue Tests of Rail Webs. R. S. Jensen. (University of Illinois, 1948, Engineering Experiment Station Bulletin, Reprint Series No. 39, pp. 21–25). Specimens of painted and unpainted rail webs were subjected to bending fatigue tests while drops of 36% H_2SO_4 were allowed to fall on them at 10 drops/min. Best results were obtained with a paint scheme consisting of one wash-priming coat of 19% solids, pigmented vinyl resin solution, two coats of a red-lead/vinyl-resin primer, and a finish coat pigmented with carbon black, red lead, and extenders.—R. A. R.

Sixth Progress Report of the Rolling-Load Tests of Joint Bars. R. S. Jensen. (University of Illinois, 1948, Engineering Experiment Station Bulletin, Reprint Series No. 39, pp. 26–35). The results of rolling-load fatigue tests on different types of joint bars used by railways in the U.S.A. are reported. R. A. R.

The Problem of Relieving Stress by Fluctuating Loads. H. Neerfeld and H. Möller. (Mitteilungen aus dem Max-Planck-Institut für Eisenforschung: Archiv für das Eisenhüttenwesen, 1949, vol. 20, May–June, pp. 205–210). The effect of fluctuating tensile stresses on the stress distribution of steel specimens under a static tensile load was investigated by X-ray determinations of the stress. Flat specimens of an alloy steel (C 0·23%, Cr 0·74%, Mo 0·40%) with notches on two opposite faces and others without notches were used. The application of fluctuating stresses with a maximum value less than the elastic limit of unnotched specimens causes a definite reduction in the stress determined at the surface by X-rays. The fatigue life of the specimens is improved by the reduction in stress induced by the residual compressive stress.—R. A. R.

Surface Finish, Hardness and Life of Steel Parts. C. Lipson. (Product Engineering, 1949, vol. 20, June, p. 179). A diagram is presented with which the fatigue life of a part can readily be calculated when the design stress, hardness, and surface finish are known. Similarly, if the hardness, surface finish, and cycles of stress the part is to withstand are known, the permissible stress can be obtained.—R. A. R.

Fatigue of Welded and Riveted Trusses. C. Cerardini. (Welding Journal, 1949, vol. 28, pp. 241-s–245-s). This paper describes static and fatigue tests carried out on welded and riveted roof trusses at the Federal Testing Laboratories at Zürich, Switzerland. The essential features of the welded trusses are: (1) The selection of cross-section of members and shape of gussets to provide continuous transmission of stress and avoid stress concentration, (2) the avoidance of notches in the members and gusset plates by the machining of joints before welding, (3) the use of extreme care in welding. The author concludes that if these features are retained in practice, welded trusses are of equal static and fatigue values with riveted trusses.—J. P. S.

Hardness Testing. J. Winning. (Mechanical World, 1949, vol. 125, June 3, pp. 641–643). Diamond and ball-indenter hardness testing is described. The advantages of the former are pointed out. The use of a clock gauge for measuring the depth of impressions led to the development of the Rockwell system which uses a precision mechanism of relatively delicate construction which should be used only on comparatively highly finished parts.—R. A. R.

Brinell Hardness as a Function of the Parameters of the Plasticity of Metals. G. P. Zaitsev. (Zavodskaya Laboratoryia, 1949, vol. 15, June, pp. 704–717). [In Russian]. A theoretical study is made of the process of plastic deformation occurring during the pressing of a sphere into a metal, the deformed region near the imprint being considered in detail. Relationships are shown to exist between the parameters of plasticity determined with the aid of an embedded sphere and those determined by stretching. The equations deduced are shown to give results in good agreement with experimental values for the plasticity parameters of a large range of steels subjected to different heat-treatments, as well as for some non-ferrous metals. Some other aspects of Brinell hardness are also considered.—S. K.

X-Ray Determination of the Distribution in Depth of Hardening and Nitriding Stresses. R. Glocker and H. Hasenmaier. (Zeitschrift für Metallkunde, 1949, vol. 40, May, pp. 182–186). An investigation of methods of improving the definition of X-ray back-reflection pictures of case-hardened and nitrided steels is reported. The use of a monochromator consisting of a pentacrythrite crystal greatly improves the definition. The stresses at different depths were determined by etching in stages and X-ray back-reflection methods. With case-hardened steel the maximum stress is at the case–core interface, with nitrided steel it is just below the surface. The stress in the latter steel is usually higher than that in the former. Increasing the ammonia dissociation from 5 to 10% considerably reduces the residual stress after nitriding.—R. A. R.

Effect of Aluminium and Boron on the Hardenability of Steel. B. E. Somin. (Stal, 1947, vol. 7, pp. 630–636 [in Russian]: Chemical Abstracts, 1949, vol. 43, May 10, col. 3330). The effect of aluminium (0·1–0·5%) was studied on chromium, manganese, silicon, and nickel steels, and of boron (0·001–0·005%) on chromium and chromium–nickel–molybdenum steels. No relation was found between the quantity of aluminium added and its effect. The effect of aluminium on hardenability varied with the kind of steel. In all cases there was an increase in the hardenability; the greatest increase was in nickel steel, the smallest was for carbon steel. The specific effect of boron added as Be_2O_3 or as FeB also depended on the steel to which it was added. Added to low-alloyed chromium–nickel–molybdenum steels, boron lowered the hardenability: with 0·0015% of boron added to chromium steel, the hardenability was not affected, 0·003% raised the hardenability. The action of boron added as an alloy with other elements (aluminium, zirconium, and titanium) was masked by the other elements.

PV Test Applied to Gearing Steel. J. G. Christ. (Iron Age, 1949, vol. 164, Aug. 11, pp. 88–90). The requirements for a gear steel are high surface hardness, heavy load-carrying capacity, and ability to stand shock loading. These can be ascertained by the 'PV hardenability test,' where a cylindrical specimen with an end cut to a 90° wedge is used. This is heated above the critical range for 1 hr., and water-quenched in a special fixture. Hardness tests are made on the surface of a slip cut longitudinally from the specimen. Typical curves for suitable steels are shown and compared with curves for steels with too high a hardenability which would harden too deeply to retain a tough core.—J. P. S.

The Portevin/Le-Chatelier Phenomenon in Mild Steels and Its Relation to the Ageing of a Cold-Worked Test-Piece. C. Boulanger. (Comptes Rendus, 1949, vol. 228, June 27, pp. 2026–2028). Portevin and Le Chatelier observed on stress/elongation curves discontinuities due to the release of waves of plastic deformation which travelled through the specimen as though it were receiving a series of successive instantaneous impacts. The present author has studied this phenomenon (which is one of the manifestations of blue-brittleness), using fine wire specimens and the Chevenard micromachine. With a 0·04% carbon steel the stress–strain curves taken at 170° and 200° C. showed very clearly the broken form of the lines, which appeared to have been 'hatched.' Increasing the rate of elongation raised the temperature at which the effect was observed. By heating the steel in hydrogen above the Ac_3 point the grain grew, and the flat part of the curve and the 'hatchings' disappeared; if the temperature in the hydrogen did not exceed 750° C. the grain remained fine, the flat remained, though it was less angular, and there was no 'hatching.' Such hydrogen-treated steel would not age after being cold-worked and heated at low temperature. The reintroduction of carbon ($C_6H_{10} + H_2$ at 725°), nitrogen ($NH_3 + H_2$ at 575°) or oxygen

($H_2O + H_2$ at 725° C.) into the steel caused the reappearance of the 'hatchings.' This treatment did not make the flat part reappear unless the grain was fine, but in all cases the metal became susceptible to ageing.—A. E. C.

Effect of Carbon and Nitrogen on Temper Brittleness. D. C. Buffum, L. D. Jaffe, and W. P. Clancy. (American Institute of Mining and Metallurgical Engineers, Technical Note No. 17: Transactions of the American Institute of Mining and Metallurgical Engineers, 1949, vol. 185, pp. 499–500: Journal of Metals, 1949, vol. 1, Aug., Section 3, pp. 499–500). A preliminary note is presented of an investigation into the effect of carbon and nitrogen on temper brittleness. A steel with 0·003% of carbon and 0·0004% of nitrogen (otherwise within the SAE 3140 composition range) was used.

Investigation of Temper Brittleness in Low-Alloy Steels. S. A. Herres and A. R. Elsea. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2574; Transactions of the American Institute of Mining and Metallurgical Engineers, 1949, vol. 185, pp. 366–370: Journal of Metals, 1949, vol. 1, June, Section 3, pp. 366–370). Twenty experimental low-alloy steels were tested for temper embrittlement by comparing notched-bar impact values at + 75° and − 40° F. of specimens water-quenched and furnace-cooled from the tempering temperature. For these conditions of testing, no embrittlement was apparent for nickel additions up to 5% to a base analysis of 0·25% of carbon, 0·2% of manganese, 0·2% of silicon, 0·02% of phosphorus, and 0·02% of sulphur. Chromium 1·5% or either 0·1% of vanadium or 0·1% of titanium with 2% of nickel gave slight embrittlement. Manganese 1·5% or 1% of manganese with 0·75% of chromium produced extreme embrittlement. Deoxidation practice or grain-size did not appear to have any direct influence on embrittlement. A moderate amount of molybdenum decreased embrittlement of the manganese–chromium steel. Increasing the phosphorus content of a 3% nickel steel produced temper brittleness, and decreasing the phosphorus content from 0·02 to 0·005% greatly decreased the embrittlement of the susceptible 1·5% manganese and 1% manganese–0·75% chromium steels. Impact-resistance, transition-temperature curves for the latter steels indicate a real influence of small differences in phosphorus content on temper-brittleness, but in the absence of fundamental data on the behaviour of phosphorus in relation to carbon and alloying elements in steel, no theory of the mechanism of this influence is apparent.

On the Sliding Properties of Cast Iron and Its Use as Bearing Material. J. Doškář. (Hutnické Listy, 1949, vol. 4, Jan., pp. 6–10; Feb., pp. 45–48; Apr., pp. 112–115). [In Czech]. For bearings where the loads are small Russian authors recommend alloy cast iron with 3·5% C, 2·5% Si, 0·65–0·85% Mn, 0·15% Cr, 0·15% Ni, 0·4% Cu, 0·12% (max.) S, and 0·2% (max.) P, and for heavy-duty bearings cast iron with 3·0–3·4% C, 1·6–1·9% Si, 0·6–0·8% Mn, 0·5% P, and 0·1% S; the structure must consist of pearlite, finely distributed phosphide eutectic, and graphite, with no cementite and less than 15% of ferrite. In making the iron the object is to prevent the formation of primary ferrite and obtain a purely pearlitic structure without any cementite, or to have most of the carbon in the combined form and then transform it into fine α-graphite by heat-treatment. Frictional properties of cast iron in segment-shaped specimens were tested in an Amsler machine, the moving part being a cylinder of hardened steel. The times for the surface temperature to reach 80–85° C. with good and with insufficient lubrication were noted. In heavy-duty bearings subject to alternating loads best results were obtained with a 4% carbon iron heat-treated to precipitate most of the carbon as graphite. Aluminium and copper additions greatly reduced the friction, but also lowered the strength. Cast iron with sliding properties equal to those of bronze can be produced in certain cases.—E. G.

The Seizure of Metals. F. P. Bowden and D. Tabor. (Institution of Mechanical Engineers, Apr., 1949, Advance Copy, pp. 1–7). The physical principles involved in the seizure of metals are discussed. In the seizure of sliding metals the local welds formed at points of intimate contact are often stronger than the metals themselves and heavy damage may occur which extends beyond the area of intimate contact. The behaviour of the metals depends on their properties at the high temperatures which obtain at the points of contact. The effects of interposing surface films of oxide, soft metals, boundary lubricants, or hard metallic films are discussed.

R. A. R.

Soft Magnetic Materials. E. A. Gaugler. (Product Engineering, 1949, vol. 20, July, pp. 84–89). The magnetic properties of magnetic materials for non-permanent magnets are presented and discussed. These include magnetic iron, silicon steel, nickel–iron alloys, iron–cobalt alloys, iron–nickel–cobalt alloys, and the mixed ferrites.—R. A. R.

A Review of the Properties of Magnetic Sheet Steel and Its Use in Engineering. D. Edmundson. (Sheet Metal Industries, 1949, vol. 26, June, pp. 1199–1204, 1214). The magnetic requirements of steel sheets used in the construction of alternators, synchronous motors, induction motors, D.C. motors and D.C. generators are considered in some detail. Brief particulars are given of a new British Thomson-Houston apparatus for testing the magnetic properties and grading separate sheets at the rate of 10/min.—R. A. R.

Nonmetallic Materials Developed with Improved Magnetic Properties. (Materials and Methods, 1949, vol. 29, June, pp. 54–55). A series of ferrites have been used for making magnets for high-frequency instruments used in communications. They include the ferrites of cupric copper, magnesium, manganese in the divalent state, nickel, ferrous iron, and zinc. They have a crystal structure like that of spinel $Mg(Al_2O_4)$. There are two variations of this, the 'normal' and the 'inverse' spinel; ferrites crystallizing in the inverse form are ferromagnetic and those taking the normal form are not. A high-permeability ferrite called 'Ferroxcube III' has been developed, the magnetic properties of which are given.—R. A. R.

Investigations of Different Pauschal Non-Magnetic States. H. Fahlenbrach and K. Sixtus. (Zeitschrift für Metallkunde, 1949, vol. 40, May, pp. 187–193). Measurements on different materials have shown that the initial permeability of specimens brought into the Pauschal non-magnetic state by demagnetizing with an alternating current is reduced after severely jolting the specimens. This phenomenon was studied using specimens of iron alloyed with nickel, silicon, and aluminium. The reduction was between 5 and 30% of the permeability after jolting and was independent of the coercive force. The initial permeability of the jolted specimen was increased by the application of a temporary alternating or direct current field provided its strength was less than the coercive force. The permeability in the jolted state was equal to that after annealing at above the Curie point.

R. A. R.

Effect of Heat-Treatment on the Magnetic Properties of a Chromium Magnet Steel. W. Jellinghaus. (Mitteilungen aus dem Max-Planck Institut für Eisenforschung: Archiv für das Eisenhüttenwesen, 1949, vol. 20, July–Aug., pp. 249–254). The relation between transformation temperature, hardening temperature, hardening time, and the magnetic properties of a chromium steel (Cr about 5%, Mn about 1%) after isothermal transformation of the austenite was investigated. Using normal hardening temperatures the properties in the pearlite stage are characterized by high remanence and low coercive force. In the intermediate stage the incomplete transformation causes lower permanent magnetization. The coercive force increases with diminishing temperature and reaches a maximum at the temperature where tempering begins. The coercive force in the centre part of the martensite stage of the steel transformed at about 150° C. is surprisingly great even after prolonged holding times. The cooling below 0° C. of specimens quenched from ordinary hardening temperatures and from about 1050° C. increases the saturation and the permanent magnetization, and slightly decreases the coercive force. The saturation and remanence of specimens quenched from 1050° C. and isothermally transformed are low and of no practical value.—R. A. R.

Reviews of Certain Aspects of Metal Physics. Part III—The Principal Work in Ferromagnetism since about 1938. W. Sucksmith. (Journal of The Iron and Steel Institute, 1949, vol. 163, Sept., pp. 51–60). This continues a series of papers (*see* Journ. I. and S.I., 1949, vol. 162, July, pp. 300–324). At the beginning of the period under review, the general outline of ferromagnetic theory had taken definite shape, whilst the relationships with other physical phenomena were gradually being built up out of the complexity of ferromagnetic effects. Rapid progress has now been made on

both the experimental and theoretical sides, whilst the gradual elimination of *ad hoc* and formal theories is taking place. The paucity of co-ordinated experimental data suitable for theoretical development, limited of necessity to the three common ferromagnetic elements, has directed a large body of work towards alloys, and it is in the field of structure-sensitive properties that major developments have taken place. New materials, both the high-permeability, low-loss type, and the high-coercivity and remanence type suitable for permanent magnets, are emerging, and the physical factors responsible for these particular properties are gradually becoming classified. The physics of the magnetization curve, with the improved conceptions of domains, has been markedly developed, largely by work on single crystals, whilst recent notable contributions to the theory of hysteresis show promise of co-ordinating the mass of data already available. In many aspects, close quantitative correlation between theory and experiment is lacking, but the gap is narrowing steadily.

Measurement of the Gyromagnetic Constant of Iron and Nickel. A. J. P. Meyer. (Comptes Rendus, 1949, vol. 228, June 20, pp. 1934–1935). The ratio of the kinetic moment, j, to the magnetic moment, μ, of an electron called the gyromagnetic ratio, is given by:

$$\rho = \frac{j}{\mu} = \frac{2mc}{ge},$$

where g is the factor of Landé, which can be measured by this relationship. Previously determined values of g being in disagreement, the author made new measurements, and found $2{\cdot}008 \pm 0{\cdot}007$ for iron and $1{\cdot}996 \pm 0{\cdot}003$ for nickel. The theoretical value is 2.—A. E. C.

Crankshaft Damping. P. Draminsky. (Institution of Mechanical Engineers, Feb., 1949, Advance Copy). The author attempts to give a correct physical explanation of natural damping by torsional vibrations, and also to obtain approximate formulæ for pre-calculation of the damping in any given case. Experimental work with a single-cylinder engine driven by external power and excited to torsional vibrations by a spring-loaded cam disc is described. A simple and practical method for the calculation of damped vibrations in arbitrary elastic systems, and of hysteresis and bearing damping in a single-cylinder engine is given.—R. A. R.

Ultrasonics and Its Technical Applications, Especially for Testing Materials. A. Leon. (Osterreichischer Maschinenmarkt und Elektrowirtschaft, 1949, vol. 4, May 1, pp. 155–160). In this review the author explains the theory of ultrasonics and describes electrical, magnetic, and gas-impulse methods of producing ultrasonic waves. Examples of its applications include the degassing of molten metals and glass, removal of dust from gases, determining elastic properties, and the detection of defects in metals.—R. A. R.

Immersed Ultrasonic Inspection. Rebecca H. Smith and D. C. Erdman. (Iron Age, 1949, vol. 164, Aug. 4, pp. 83–88). In immersed ultrasonic inspection the quartz crystal and the specimen, or part of it, are immersed in water, or other liquid to minimize multiple reflections; other advantages gained from the procedure are that surfaces need not be so smooth, or areas so flat, as with contact arrangement.—J. P. S.

Ultrasonic Testing of Aircraft Components. W. C. Hitt. (Iron Age, 1949, vol. 163, June 23, pp. 66–70). The use of the "Sonizon," an ultrasonic testing apparatus made by the Magnaflux Corp., of Chicago, is described. Examples given are the location of thin places in drawn sheet, laminations in rolled plate and the testing of the bond between silver bearing-material and its steel backing.—J. P. S.

Ultrasonic Testing of Tool Steels. J. C. Hartley and E. K. Mull. (Iron Age, 1949, vol. 163, May 19, pp. 80–85). From the examination of a number of samples of high-speed steel, the authors conclude that the Sperry Reflectoscope, an ultrasonic device, will reveal both mechanical defects, such as hot tears, and segregation.—J. P. S.

Examination of Materials by the Sound-Vision Method. R. Pohlman. (Iron and Steel Institute, 1949, Translation Series, No. 383). This is an English translation of a paper which appeared in Die Technik, 1948, vol. 3, Nov., pp. 465–470. *See* Journ. I. and S.I., 1949, vol. 162, May, p. 121.

Industrial Uses of Radioisotopes. (Iron Age, 1949, vol. 163, June 9, pp. 60–61). This is a bibliography, prepared by Arthur D. Little, Inc., on the use of radioactive isotopes in industry, which includes references to 27 articles on their employment in metallurgy, in friction studies, and in radiography, and to a number of patent specifications referring to the preparation of suitable isotopes.—J. P. S.

X-Ray Methods. J. F. Hinsley. (British Iron and Steel Research Association, Conference on the Non-Destructive Testing of Steel Castings, Sept., 1948, pp. 10–14). Methods of obtaining radiographs of castings are described, and the principal terms used in X-ray work are explained by simple analogies or illustrations.—R. A. R.

A Nomogram for Computing Geometrical Unsharpness in Industrial Radiographs. H. E. Seemann. (Non Destructive Testing, 1949, vol. 7, Spring Issue, pp. 12–15). Since screen and film qualities introduce an appreciable amount of unsharpness in radiographs, it is not worth while reducing the geometrical unsharpness beyond the point where it ceases to be noticeable. Tests are described for determining the upper limit for geometrical unsharpness and a nomogram is presented for computing it.—R. A. R.

Gamma Radiography. R. Halmshaw. (British Iron and Steel Research Association, Conference on the Non-Destructive Testing of Steel Castings, Sept., 1948, pp. 15–19). The application of gamma radiography for examining steel castings and its advantages and disadvantages as compared with X-rays are briefly stated.—R. A. R.

X-Ray Apparatus for Industrial Use. J. C. Rockley. (British Iron and Steel Research Association, Conference on the Non-Destructive Testing of Steel Castings, Sept., 1948, pp. 20–24). A general review of present-day X-ray apparatus suitable for inspecting steel castings and welded parts, with a brief idea of the capabilities of the different types of equipment, is presented in a manner which provides assistance to the industrialist in choosing equipment most suited to his requirements.—R. A. R.

Radioactive Materials. W. P. Grove. (British Iron and Steel Research Association, Conference on the Non-Destructive Testing of Steel Castings, Sept., 1948, pp. 25–27). The cost and the supply position regarding radioactive materials in Great Britain are discussed. The Ministry of Supply has established the Radiochemical Centre at Amersham, the function of which is to provide radioactive materials for anybody who wants them. The present rental charged by the Ministry for a 0·25 g. source of radium is about £55 a year. At the end of 1949 the Centre expects to be able to provide radon for all who need it for radiography. At present there are available in Great Britain only a few curies of mesothorium; for experimental purposes it costs about £70 per millicurie. Radioactive cobalt with a half-life of 5·3 years is a product readily obtainable in very high activities from the atomic pile. When uranium undergoes fission in the atomic pile the atoms are split into two parts and each of these is radioactive so that this may develop into a source of products useful for gamma radiography.—R. A. R.

Radiology Abroad. A. R. Greatbach. (British Iron and Steel Research Association, Conference on the Non-Destructive Testing of Steel Castings, Sept., 1948, pp. 28–30). The author reports his impressions of the state of development of industrial radiology abroad, mainly in America, and makes some comparisons with conditions in Great Britain.—R. A. R.

Magnetic Testing Methods. P. Rowbotham. (British Iron and Steel Research Association, Conference on the Non-Destructive Testing of Steel Castings, Sept., 1948, pp. 33–42). The principles of electromagnetic flaw detection are explained and various forms of apparatus are described and illustrated. R. A. R.

Ultrasonic Testing Methods. D. O. Sproule. (British Iron and Steel Research Association, Conference on the Non-Destructive Testing of Steel Castings, Sept., 1948, pp. 43–50). The principles used and the apparatus for applying ultrasonic testing methods to steel are described and the advantages and limitations of the method are pointed out.—R. A. R.

An Industrial Application of Ultrasonic Testing. G. T. Harris. (British Iron and Steel Research Association, Conference on the Non-Destructive Testing of Steel Castings, Sept., 1949, pp. 51–54). The particular application of ultrasonic testing to forged steel turbine discs is described. The dimensions and shape of the piece make it more suitable for testing by ultrasonics than by X-rays or gamma rays. The examination of the ingot, bloom, and disc by this method are described. Experience proved the success of the use of the new instrument to pass or reject many thousands of very expensive forgings.—R. A. R.

Recent Work on a Comparison of X-Ray, Gamma-Ray and Ultrasonic Methods. G. M. Michie. (British Iron and Steel Research Association, Conference on the Non-Destructive Testing of Steel Castings, Sept., 1948, pp. 55–60). Two series of investigations are described : one to compare radiographic and ultrasonic testing methods, and the other to compare radiographic methods. It is concluded that in the examination of simple parallel-sided sections, the ultrasonic method is capable of revealing centre-line shrinkage in a manner not inconsistent with radiographic evidence, but it is questionable if the method can be relied on to detect small flaws underlying a rapid change in casting thickness. The second series of investigations has not proceeded far enough for any conclusions to be reached.—R. A. R.

Radiographic Inspection Standards. W. J. Wiltshire. (British Iron and Steel Research Association, Conference on the Non-Destructive Testing of Steel Castings, Sept., 1948, pp. 61–63). An explanation is given of what is involved in setting up standards for radiographic inspection. Radiographic inspection comprises three stages. The first is the taking of the radiographs, and an inspection standard must include a specification of the technique to be used. The second, which is incapable of standardization, is the interpretation of the radiographs. The third stage is deciding what is to happen to the casting in the light of the radiographic findings ; this is entirely an engineering problem.—R. A. R.

Heat-Resisting Cast Iron. G. Clas and K. Houben. (Neue Giesserei, 1949, vol. 36, May, pp. 131–138). The physical-chemical aspects of the changes in cast iron on heating and the stabilizing effect of certain alloying elements are examined. A method of calculating the growth of cast iron caused by the dissociation of iron carbide is demonstrated. Extensive tests on the heat resistance of high-silicon iron tubes, some alloyed with 5·5% of aluminium, others with up to 2% of chromium, and a third lot unalloyed, in an air preheater have been carried out as well as laboratory tests indicating the weight of scale formed and the increases in weight and volume. The poor resistance of chromium cast iron in blast-furnace gas at 450–550° C. and the favourable effect of even 1% of aluminium on the chemical stability at 700° C. were observed ; the detrimental effect of silicon and aluminium on the mechanical properties must, however, be taken into account.—R. A. R.

Cyclic Heating Test of Main Steam Piping Joints between Ferritic and Austenitic Steels—Sewaren Generating Station. H. Weisberg. (Transactions of the American Society of Mechanical Engineers, 1949, vol. 71, Aug., pp. 643–664). A cyclic heating test of several full-size thick-walled welded pipe joints between austenitic and ferritic steels is described. One hundred cycles of alternate heating and cooling of several full-size pipe joints between austenitic and ferritic steels through the range of temperature expected in normal operation produced no apparent deleterious results. Although it is agreed that long-time exposure at the operating pressure and temperature cannot be evaluated in a short-time test, it is believed that the results provide considerable assurance that 18/8 steel piping and particularly welds between this piping and low-chromium still piping, can satisfactorily withstand the temperature changes which may be expected in power-plant service. A lengthy discussion on this paper is appended and includes details of work carried out by other investigators.—J. C. R.

The Rupture-Test Characteristics of Heat-Resistant Sheet Alloys. J. W. Freeman, E. E. Reynolds, and A. E. White. (National Advisory Committee for Aeronautics, 1948, Technical Note 1465 : [Abstract] Nickel Bulletin, 1949, vol. 22, June, p. 100). This is a report of an investigation of the properties of a representative group of heat-resisting alloys in sheet form, made with the primary object of providing data which can be used by the aircraft industry to determine whether the high-temperature rupture-test properties of a material provide a satisfactory criterion of service performance in combustion chambers and other applications of sheet alloys in power plants for aircraft. The materials studied included the standard chromium–nickel–iron alloys (35/15 nickel–chromium, 25/20 chromium–nickel, with and without silicon addition, 25/12 chromium–nickel and Inconel alloys), Vitallium, cobalt–chromium–nickel alloy (55–30–15%), S-816, S-590, low-carbon N-155, and four experimental alloys containing cobalt, molybdenum, tungsten, and boron, in addition to nickel and chromium. At 1700° F. (925° C.) the degree of superiority shown by the more highly alloyed materials, over the standard nickel–chromium–iron alloys, decreases with time to rupture ; at time periods of over 1000 hr. the differences are relatively small. At 1800° F. (980° C.) the time period for near equalization is considerably shorter than at the lower temperature. For most of the time periods considered, sheet Vitallium, S-590, low-carbon N-155, and one of the experimental complex alloys have the highest strength values. Of the standard materials, the 25/20 chromium–nickel–iron showed the best behaviour. Variations in behaviour among different samples of the latter alloy and of Inconel suggest that prior heat-treatment has an important influence on high-temperature properties. This conclusion is substantiated by the results of other investigations. Oxidation conditions during test also influenced the behaviour of the alloys, indicating that corrosive conditions during service might have an important bearing on useful life. The tendency towards equalization of rupture strengths, irrespective of alloy content, is considered to be due to agglomeration, solution, or lack of precipitation of the excess constituent believed to control strength at lower temperatures.

Boron–Titanium Steels for Moderate Temperatures. G. F. Comstock. (Iron Age, 1949, vol. 163, June 16, pp. 90–92). A series of low-carbon 18/11 chromium–nickel steels (Mn 0·19–1·29%), but all containing 0·2–0·4% of titanium was made and to each was added 0·02–0·03% of boron. Tests for air-cooled hardness, quenched hardness, room-temperature tensile and impact tests, and stress-rupture tests at 1000° F., 1100° F., and 1200° F. (540° C., 595° C., and 650° C.) were performed. The conclusions are : (1) A plain carbon steel gives greater hardness, both air-cooled and water-quenched, when containing boron and titanium than when containing either element alone ; (2) similar steels containing boron and titanium have higher room-temperature tensile properties when air-cooled and tempered, with considerable loss in elongation but little in reduction of area, compared with the air-cooled and untempered condition ; (3) titanium–boron-containing carbon steels have greater elevated-temperature stress-resisting capacities than the molybdenum–titanium and carbon–molybdenum steels in the series ; (4) a 3% chromium steel containing boron and titanium has a higher rupture strength at 1100° F. (595° C.) than that reported for any other ferritic steel ; (5) similar improvements in properties are made to 6% chromium steels, to 18% chromium, 11% nickel steels, and to molybdenum steels by the addition of boron and titanium ; and (6) the presence of titanium inhibits graphitization at high temperatures.—J. P. S.

Low-Alloy Hot-Working Steels. H. Krainer, K. Swoboda, and F. Rapatz. (Archiv für das Eisenhüttenwesen, 1949, vol. 20, Mar.–Apr., pp. 111–114). With a view to saving tungsten and molybdenum, tensile and hot-cracking sensitivity tests were made on 40 low-alloy chromium steels with and without additions of vanadium, manganese, silicon, and small amounts of molybdenum. Vanadium makes a marked contribution to the stability of steels at high temperatures, but with high vanadium contents, the chromium must be kept at about 1·5%. Molybdenum, besides improving the deep-hardening properties, greatly increases the hot strength. With the lowest possible additions of tungsten and molybdenum, the best results were obtained with the following composition : Cr 1·5%, Mn 0·5%, V 0·9% and W 0·5%.

R. A. R.

The Strength Behaviour of Low-Alloy Heat-Resisting Steels and their Tendency to Brittle Fracture. K. Richard. (Archiv für Metallkunde, 1949, vol. 3, May, pp. 157–164). Methods of determining the strength of steels at high temperatures, the causes of brittle fracture, the effect of notches, damage preceding fracture, preventing embrittlement by heat-treatment, and the strength of tubes and bolts of heat-resisting steels are discussed. It is shown that data from 45-hr. tests cannot be used to predict the long-time creep strength of heat-resisting steels and that low-alloy steels on the whole are unable to sustain loads for a long time at temperatures above 450° C.—R. A. R.

Creep in Metals and Methods of Creep Testing. M. Randall. (Machinery, 1949, vol. 74, June 9, pp. 772–773). Creep and creep-testing methods are discussed. The normal type of creep curve can be divided into three parts. The first stage is one of rapid extension (primary creep), the second stage is one of slow extension (secondary creep), and the third is again one of rapid extension (tertiary creep). The minimum

creep rate is determined from the second stage and fracture can only occur in the third stage.—R. F. F.

Selection of Heat Resistant Steels. J. B. Henry. (Product Engineering, 1949, vol. 20, July, pp. 113–118). Data are presented on tensile and fatigue strengths at high temperatures and the resistance to scaling of a wide variety of high alloy chromium and chromium–nickel steels.—R. A. R.

Behaviour of Cast Steel at Elevated Temperatures. A. E. Johnson. (Engineer, 1949, vol. 188, July 29, pp. 126–128; Aug. 5, pp. 138–141; Aug. 12, pp. 165–168; Aug. 19, pp. 189–193). About 80 tests consisting of pure tensile, pure torsion, and complex stress creep tests have been performed on 0·17% carbon cast steel at temperatures of 350°, 450°, and 550° C. The results are analysed and equations for the stress–strain relationships are derived. The following are some of the conclusions reached: (1) The behaviour of the steel was essentially isotropic at all except the highest stresses in the tests at 350° C. At 450° C. relatively simple anisotropy was shown at all except the highest stresses; at these latter stresses the anisotropy occurring was, as at 350° C., apparently due to high initial plastic strain. At 550° C. a very complex condition of anisotropy occurred. It is possible that the anisotropy exhibited at 450° and 550° C. is associated with changes occurring in the ferrite–pearlite structure of the material at these temperatures conferring directional creep properties; and (2) At 350° C. the octahedral shear strain appeared for all stress systems to be a definite function of the octahedral shear stress, *i.e.*, the Hencky criterion of plastic strain appeared to be fulfilled approximately for the material.—R. A. R.

The Behaviour of Steel at Elevated Temperatures. A. Pomp. (Stahl und Eisen, 1949, vol. 69, Apr. 14, pp. 270–273: Apr. 28, pp. 310–313; May 12, pp. 339–342). This is a review of the literature for the years 1944 to 1947.—R. A. R.

Recovery in Creep Testing. A. Krisch and A. Pomp. (Mitteilungen aus dem Max-Planck-Institut für Eisenforschung: Archiv für das Eisenhüttenwesen, 1949, vol. 20, May–June, pp. 189–195). When specimens have been pulled beyond the elastic limit there is some recovery when the load is removed; this recovery does not cease suddenly, but continues at a decreasing rate for some time. This was studied using two mild steels (C 0·03% and 0·23%) and an austenitic steel (Cr 15%, Ni 13%, W 2·5%) subjected to creep tests. The recovery increased with increasing temperature and load in the tests. After changes in the duration of the creep tests, maximum recovery was observed after 48 hr. under load. The time elapsing until recovery ceases does not increase directly with the amount of recovery.—R. A. R.

The Relaxation Tests in Terms of Creep and Creep Recovery. E. A. Johnson. (Metallurgia, 1949, vol. 39, Apr., pp. 291–297). The author attempted to ascertain the relative importance of creep recovery as a link between the relaxation test (one in which the total strain incurred upon applying the initial load is subsequently kept constant by reduction of the current load) and normal creep tests. Relaxation tests from relatively high and relatively low initial total strains have been made on a 0·17% carbon steel at 445° C., and on a chromium–molybdenum steel at 485° C. The correction to be applied to observed creep rates for effects of creep recovery is only of order 7% for the former steel and of order 11% for the latter.—R. A. R.

The Properties of Metallic Materials in the Stratosphere. P. L. Teed. (Technique et Science Aéronautiques, 1947, No. 6, pp. 341–359). The behaviour of metals and alloys used in the aircraft industry, under the conditions prevalent in the stratosphere, is discussed. Due to the dryness, low density, and low temperature of the stratosphere, the dangers of corrosion and stress corrosion are less than at ground level. The mechanical properties of most metals are improved at these low temperatures, though Invar steel shows a decrease in elasticity, and tin and certain alloy steels have a lower tensile strength and elastic limit. The changes in physical properties, due to the lower temperature, are reversible for most metals, tin being the main exception. A bibliography of 24 references is appended.—R. F. F.

Cast Ferrous Materials for Sub-Zero Service. J. W. Juppenlatz. (Iron Age, 1949, vol. 163, June 9, pp. 46–49). Plain carbon steels retain adequate impact properties down to 0° F. (− 18° C.), ferritic low-alloy steels are serviceable down to − 75° F. (− 57·2° C.) and, when properly heat-treated, down to − 150° F. (− 101° C.); austenitic stainless steel can be used down to temperatures of − 425° F. (− 254° C.), and is thus suitable for handling liquid oxygen. J. P. S.

Equilibrium in the Reaction of Hydrogen with Oxygen in Liquid Iron. M. N. Dastur and J. Chipman. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2661: Transactions of the American Institute of Mining and Metallurgical Engineers, 1949, vol. 185, pp. 441–445: Journal of Metals, 1949, vol. 1, Aug., Section 3, pp. 441–445). An experimental study has been made of equilibrium in the reaction of hydrogen with oxygen in liquid iron in the temperature range 1563–1760° C. The primary purpose was to secure more dependable data by elimination of certain errors which were inherent in earlier work. Errors of thermal diffusion in the gas mixture were eliminated by preheating the entering gas stream to the temperature of the metal. A further safeguard was the admixture of pure argon with the gas stream in a ratio of about 4 to 1. Temperature measurements were made by an optical pyrometer sighted on the clean surface of the melt. This system was calibrated in the range 1530–1930° C. against optical and thermoelectric measurements in a re-entrant tube in the bottom of the crucible. This reduced uncertainties in temperature measurement to about ± 5°. The experimental results are represented by

$$K_1 = P_{H_2O}/P_{H_2} \times \%\,\underline{O}$$

and:

$$\log K_1 = 7050/T - 3{\cdot}17$$

where $\underline{O}$ denotes oxygen in liquid iron.

Related thermodynamic quantities are derived. The results confirm the oxygen solubility data of Fetters, Taylor, and Chipman and are in good agreement with equilibrium data on the system iron–oxygen of Darken and Gurry.

Nitrogen in Steel. S. Feigenbaum and G. H. Enzian. (Iron Age, 1949, vol. 163, June 30, pp. 52–54). The authors discuss the history of the study of the effect of nitrogen in steel. While its rôle appears to be fairly well explained, the development of chemical methods for isolating and identifying the various nitrogen compounds in steel is of very recent origin and should prove a very valuable metallurgical tool. J. P. S.

Effect of Nitrogen on the Structure and Properties of Steel EI 319. N. V. Semenova. (Stal, 1947, vol. 7, pp. 529–533 [in Russian]: Chemical Abstracts, 1949, vol. 43, Apr. 25, cols. 2912–2913). This chromium–nickel steel is used frequently for welding electrodes. Ordinarily it contains ferrite and austenite, a fact which, because of the different plastic properties, causes 'culls' in forging and rolling. Experiments were carried out to determine the effect of nitrogen on depression of ferrite formation. The presence of nitrogen arrested ferrite formation. At a nitrogen/chromium ratio 1 : 60, the structure of the steel consisted of austenite. At a ratio of 1 : 100 there appeared a component of eutectic nature. At smaller nitrogen/chromium ratios, ferrite was present. The mechanical properties of this steel containing 0·26–0·27% nitrogen after thermal treatment are the same as of steel without nitrogen. Welds made with electrodes from this nitrogen-bearing steel on armour plate were free of defects. The weld contained less nitrogen than the electrode.

Study of the Reaction between Solid Oxides under Vacuum. H. Forestier and J. P. Kiehl. (Comptes Rendus, 1949, vol. 229, July 18, pp. 197–199). In an earlier research (Comptes Rendus, 1949, vol. 228, p. 47) a marked reduction in the rate of reaction between mixtures of solid oxides *in vacuo* was observed. This was followed up with tests on a mixture of Fe_2O_3 and NiO (reacting to produce Fe_2O_3NiO); it was found that the rate of formation of the ferrite was a linear function of the logarithm of the pressure and tended towards zero at zero pressure. The small amount of ferrite formed at the lowest vacuum was attributed to reaction due to a small quantity of gas still absorbed at the annealing temperature and reaction due to a feeble diffusion. Tests with co-precipitated oxides showed an enhanced influence of adsorbed gases and faster rate of reaction.—A. E. C.

The Effects of Molybdenum and Commercial Ranges of Phosphorus upon the Toughness of 0·40% Carbon Chromium Steels. M. Baeyertz, W. F. Craig, jun., and J. P. Sheehan. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2654: Transactions of the American

Institute of Mining and Metallurgical Engineers, 1949, vol. 185, pp. 535–543 : Journal of Metals, 1949, vol. 1, Aug., pp. 535–543). Using AISI/SAE 5140 steel the authors have studied the loss of toughness caused by increasing the phosphorus content from 0·02 to 0·04%, and have endeavoured to counteract this loss by replacing part of the chromium content by molybdenum. Data obtained are presented in graphical form and the following conclusions are reached : (1) Phosphorus reduces the toughness of tempered martensite even in the absence of temper brittleness. In 5140 steel this loss in toughness can be partially compensated by replacement of part of the chromium by molybdenum with no essential change in the hardenability of steel. (2) Confirmation has been obtained that phosphorus accentuates temper embrittlement of martensite. (3) Replacement by molybdenum of part of the chromium materially reduced susceptibility to loss of toughness by temper embrittlement. (4) A study of 5140 steels and molybdenum modifications of similar hardenability showed that replacement of part of the chromium by molybdenum can be used to improve the toughness at phosphorus levels encountered in commercial steels.—J. C. R.

Effect of Boron on the Structure and Some Physical Properties of Plain Cast Iron. A. I. Krynitsky and H. Stern. (Journal of Research of the National Bureau of Standards, 1949, vol. 42, May, pp. 465–479). The effects of boron additions in the 0·001–0·48% range on the properties of grey cast irons were determined by casting chill plates, wedges, and arbitration bars. Cooling curves indicated that, during solidification, boron produced an undercooling effect which increased with increasing boron content. Boron increased the depth of chill in plates and wedges as well as the hardness of arbitration bars and chill plates. It diminished the size and quantity of graphite, increased the free carbide content, and tended to increase the amount of eutectic structure. With 0·29% or more of boron, the structure of the irons consisted essentially of a matrix of cementite containing islands of pearlite and islands with a practically unresolved structure which were probably very fine pearlite. R. A. R.

The Effect of Impurities on the Properties of Metals. C. H. Desch. (Bulletin of the Institution of Mining and Metallurgy, 1949, Aug., No. 513, pp. 17–28). Metals are very sensitive to the influence of foreign atoms in their structure. If those atoms be scattered at random throughout the crystal lattice, their effect is to increase the hardness and to diminish the ductility and electrical conductivity. If, however, the foreign atoms be rejected by the crystal lattice of the solvent metal, their influence on certain properties may be far greater and less predictable. Examples of this are cited and methods of producing pure, or almost pure, metals are briefly described. It has been proved that dissolved foreign atoms lower the surface tension of molten metals in a number of instances. It is less easy to prove that a similar lowering of surface tension is produced in a solid metal, but measurements have been made which provide evidence that such a concentration of impurities in grain boundaries, although below the concentration corresponding with solubility in the mass, actually does occur.—R. A. R.

Improving Bolt Performance. A. R. Anderson. (Product Engineering, 1949, vol. 20, May, pp. 109–111). Stress distribution patterns of the stresses in bolts and nuts by the photoelastic technique have shown how the design can be improved. Tapering the shank into the head reduces the concentration of stress at the shoulder, and providing an annular groove three or four threads deep in the nut permits that part of the nut to stretch with the bolt, thus alleviating the compressive stresses.—R. A. R.

Nodular Graphite Cast Iron as an Engineering Material. (Materials and Methods, 1949, vol. 29, May, pp. 45–48). British and American methods on the production of nodular cast iron are briefly reviewed. British research has developed the method involving a small addition of cerium to the ladle shortly before casting, as described by H. Morrogh and J. W. Grant. American work, reported by C. K. Donoho, requires the addition of a magnesium alloy. Nodular cast iron can be arc-welded and the graphite remains in the nodular form. In the cast condition, and up to 300 and 320 Brinell hardness, it is as free-machining as flake graphite iron of the same hardness, but superior to grey irons of equivalent tensile strength.—R. A. R.

Gear Materials. E. B. Richardson. (Australian Institute of Metals : Australasian Engineer, 1949, May 7, pp. 43–45). The stresses to which gears are submitted and data on the maximum size of section and surface, and bending stresses applicable to teeth made of 0·4 and 0·55% carbon steels and of steels alloyed with nickel, nickel and chromium, and nickel chromium and molybdenum, are discussed.—R. A. R.

Dynamic Torsion of Metals and Alloys Used in Aircraft Construction—Their Elastic Limit and Micro-Plastic Deformation under Reversed Torsional Energy Loads. G. Welter. (Metallurgia, 1949, vol. 39, Feb., pp. 188–190 ; Mar., pp. 253–256 ; Apr., pp. 313–315). Investigations of the maximum possible impact loads causing no permanent micro-deformation of the specimen under plus and minus torsion loads are reported. Part I describes a special torsional impact testing machine for applying repeated and increasing dynamic loads. The test results on a 0·20% carbon steel, a nickel–chromium steel, Monel metal, Duralumin, and a magnesium alloy are presented and discussed in Parts II and III. Static torsional loads, and superimposed torsional impact loads in either direction could be applied. With mild steel under dynamic torsion loads of 1 lb. with increasing dropping height, about 20 in.lb./cu. in. were necessary to produce 0·0001 in./in. permanent deformation on non-preloaded specimens and 42–43 in.lb./cu. in. to produce 0·002 in./in. permanent deformation. With preloading in torsion causing 0·013 in./in. deformation, a small dynamic load in the opposite direction caused very great permanent deformation. This phenomenon (the Bauschinger effect) was observed in all the material and was greatest with mild steel. With nickel-chromium steel permanent deformations of 0·0001 and 0·0007 in./in. were produced by impacts of about 30 and 60 in.lb./cu. in. respectively. When dynamically tested in the opposite direction loads of 3, 9, and 40 in.lb./cu. in. produced permanent deformations of 0·0001, 0·00025, and 0·001 in./in. respectively.—R. A. R.

A Metallurgical Investigation of a Contour-Forged Disc of EME Alloy. E. E. Reynolds, J. W. Freeman, and A. E. White. (National Advisory Committee for Aeronautics, 1948, Nov., Technical Note No. 1534 : [Abstract] Metals Review, 1949, vol. 22, Feb., p. 22). Properties of EME alloy (19% chromium, 12% nickel, 3% tungsten, 1% columbium, iron base) in the form of contour-forged discs for the rotors of gas turbines were studied at room temperature and 1200° F. Results are compared with data from other laboratories.

Influence of Fine Cracks on the Mechanical Properties of Stainless Chromium Steel. V. I. Smirnov and N. S. Orlova. (Kotloturbostroenie, 1948, July–Aug., pp. 21–23 [in Russian] : [Abstract] Metals Review, 1949, vol. 22, Feb., p. 22). Results of experiments show that, in the case of coincidence of the direction of fine cracks with the direction of prevailing stresses, the influence is insignificant. If the directions do not coincide, a considerable decrease in plasticity is observed. Fine cracks do not decrease the corrosion resistance of stainless chromium steel in supersaturated steam.

Pearlitic Structure Effect on Brittle Transition Temperature. N. Grossman. (Welding Journal, 1949, vol. 28, June, pp. 265-s–268-s). A plain, low-carbon, fine-grained steel (0·17% Cr, 0·48% Mn) was heat-treated to obtain four different structures : (*a*) large grain, coarse pearlite, (*b*) large grain, fine pearlite, (*c*) small grain, coarse pearlite, and (*d*) small grain, fine pearlite. Transition temperatures were then determined by a simple notched bend test, this being the highest temperature at which ductile behaviour, as indicated by deflection, ceases. It was found that the smaller grain-size gives a lower transition temperature, and so does the finer pearlite, the magnitude of the changes being of the same order ; it is not clear whether the changes are due to pearlite spacing and grain-size, or to a combination of these factors as well as to the amount of free ferrite and carbide in the pearlite.—J. P. S.

New Apparatus for the Mechanical Testing of Locomotive and Other Piston Rings. P. G. Korolev. (Zavodskaya Laboratoriya, 1949, vol. 15, June, pp. 754–756). [In Russian]. An improved form of apparatus for the testing of large piston rings is described. The diameters of rings which can be tested in it range from 350 to 200 mm.—S. K.

Recent Developments in Titanium and Titanium Alloys. (Steel, 1949, vol. 124, June 20, pp. 100–104, 132–135 ; June 27, pp. 58–61, 92, 94). The first part of this review deals with

the properties of pure titanium, and with methods for its production. In the second part, some of the properties of titanium–boron steels containing 0·034% C, 0·026% B, and 0·30% Ti are described, as well as those of similar alloys containing chromium or molybdenum in addition. High-temperature strength is the main feature of these steels. Some titanium base alloys containing tungsten and nitrogen or carbon are also briefly described.—J. P. S.

METALLOGRAPHY

The Development of an Instrument for the Electrolytic Polishing of Metallographic Specimens in the Industrial Laboratory. W. Engelhardt and R. Neuf. (Archiv für Metallkunde, 1949, vol. 3, May, pp. 180–185). The electrolytic polishing apparatus described consists essentially of a stand holding a glass vessel provided with a stopper at top and bottom, and an opening in the side against which the specimen is pressed by a spring. A quartz tube closed at the upper end, projects through the lower stopper; it is wound with platinum strip to act as cathode, and is provided with cooling water internally. The upper stopper of the glass tube carries a thermometer and a vent tube, and a stirring device driven by a motor carried by the same stand. The whole apparatus can be tilted on the stand so that, on pouring the electrolyte into the glass tube, the specimen remains uncovered until the apparatus is brought to the upright position. The results of tests on aluminium and its alloys are given; these were successful and proved that the instrument was efficient and that polishing times for these alloys need not exceed 2 min.—R. A. R.

A New Etching Reagent for Cast Iron and Steel. O. Schaaber. (Neue Giesserei, 1949, vol. 36, May, p. 154). An etching solution for cast iron and steel structures consists of a 1–3% solution of concentrated nitric acid in concentrated acetic acid. The specimen, ground and polished in the ordinary way, is washed with water, then in alcohol, dried in hot air, dipped in the etching solution for 5 to 30 sec. at room temperature, washed in running water, then in alcohol, and dried. It can be re-etched without repolishing.—R. A. R.

The Phase-Contrast Microscope with Particular Reference to Vertical Incident Illumination. E. W. Taylor. (Journal of the Royal Microscopical Society, 1949, vol. 69, May, pp. 49–52). With a properly designed phase-contrast microscope it is possible to explore the infra-red and ultra-violet regions of the spectrum and to record the results on the photographic plate. A metallurgical microscope of this type is described and some micrographs of malleable iron, grey iron, tool steel, case-hardened steel, zinc, and aluminium are reproduced.—R. A. R.

Interpretation of Electron Micrographs, Prepared by the Plastic Replica Process. J. Trotter. (Nature, 1949, vol. 164, Aug. 6, pp. 227–228). With replicas of etched metal surfaces for the purpose of electron-microscope examination, it is generally assumed that the upper surface, remote from that to be examined, is plane. That this is not the case has been proved by depositing chromium on such a surface, when a considerable amount of the detail of the original metal surface is revealed. The effect of this unevenness is to introduce false effects when the side which has been in contact with the specimen is examined. It is concluded that, to ensure the perception of all the detail on this side, it should be coated, by the shadow technique, with a thick layer of a heavy metal such as chromium. This not only increases the contrasts, but eliminates the influence of the non-planar, other surface of the replica.—J. P. S.

The Study of Grain Boundaries with the Electron Microscope. J. F. Radavich. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2623; Transactions of the American Institute of Mining and Metallurgical Engi..eers, 1949, vol. 185, pp. 395–398; Journal of Metals, 1949, vol. 1, July, Section 3, pp. 395–398). Three specimens of ingot iron were used in this investigation. The first, termed 'brittle,' would not withstand a 90° bend, the second, termed 'good,' could withstand a 90° bend, and the third, termed 'cured,' withstood a 90° bend after annealing in a hydrogen atmosphere. After being polished the specimens were examined under the electron microscope to determine the location of impurities. The procedure is described and the results obtained are illustrated by photomicrographs. The evidence suggested that the cause of embrittlement might be segregation of impurities at the grain boundaries. The embrittling material was probably in the form of oxides.—J. C. R.

Note on As-Cast Structure and Grain Size in Cast Alloy Steels. E. A. Loria. (American Foundrymen's Society, May, 1949, Preprint No. 44). An etching reagent and technique have been developed to illustrate the as-cast austenite grain-size in cast alloy steels which do not develop ferrite outlining of the austenite grains formed at a very high temperature during or shortly after the freezing of the steel casting. Results obtained on four cast steels revealed large polygonal grains and the dendritic pattern developed on solidification. The rate of chemical attack of the reagent varied with the orientation of the initial austenite grains (or rather the products of their subsequent decomposition) and produced a typical grain-contrast etch. The number of as-cast grains in the central areas of etched cross-sections for each steel were counted and the average grain area compared with the A.S.T.M. grain size developed at 1550° and 1700° F. No apparent relationship was found between the original and subsequent grain-sizes.

Some Experiences in the Application of the Electron Microscope to the Study of Steels. F. W. Cuckow and J. Trotter. (Proceedings of the Physical Society, Section B, 1949, vol. 62, June 1, pp. 360–365). The 'unresolved pearlite' of an end-quenched low-alloy chromium steel, and the intermediate transformation products of a low-alloy nickel–chromium steel quenched in a lead bath, have been examined by the electron microscope, using both plastic and silica film replicas. Low magnifications (1000–1800) were used to obtain a large field. It is found that resolution is far superior to that of the light microscope, and that, though picture sharpness is somewhat greater with the silica replica, the plastic replica gives a more straightforward interpretation of surface geometry.—J. P. S.

Electron Microscope Study of Quenched and Tempered Steel. J. Trotter and D. McLean. (Journal of The Iron and Steel Institute, 1949, vol. 163, Sept., pp. 9–13). The electron microscope has been used to follow the changes occurring during the tempering of a 0·6% carbon steel quenched to martensite. Separate specimens were used for each tempering temperature and the series was examined with three different etching reagents: 1% of nitric acid in alcohol, ethereal picric acid plus a wetting agent, and chromic acid used electrolytically. The electron microscope gave five to ten times the resolution of the optical microscope, and appeared to resolve nearly all the detail revealed by etching. The structures are interpreted to mean that nuclei of enriched carbon content, 100–400 atoms in dia., form up to about 200° C., when a precipitate of thin plates about 1000 atoms ($\frac{1}{4}$ micron) in dia. commences to form. They are replaced by elongated particles about $\frac{1}{2}$ micron long at 350° C. The particles become more rounded at 550° C., and at still higher temperatures grow considerably larger. At 450–700° C., cementite is found by X-ray examination. The ferrite grain size seems to be determined by the fineness of the martensite grains into which each austenite crystal splits, and remains at about 2 microns even after tempering at 700° C., there being no obvious grain growth at this temperature. A specimen air-cooled from 700° C. contained a film along practically every grain boundary, possibly of cementite.

Microscopical Studies on The Iron–Nickel–Aluminium System. Part I—$\alpha + \beta$ Alloys and Isothermal Sections of the Phase Equilibrium Diagram. A. J. Bradley. (Journal of The Iron and Steel Institute, 1949, vol. 163, Sept., pp. 19–30). Koster's miscibility gap between the α, face-centred cubic phase, and the β, body-centred cubic phase fields of the Fe–Ni–Al system is now found to consist of five separate regions, and not merely three, as Bradley and Taylor had previously thought as the result of X-ray work only. The ordered phase fields, α' (Ni_3Al) and β' (NiAl), account for the existence of three separate duplex regions, $\alpha + \beta$, $\alpha + \beta'$ and $\alpha' + \beta'$, separated by the three-phase regions $\alpha + \beta + \beta'$ and $\alpha + \alpha' + \beta'$, respectively. At the melting point, the three-phase regions are extremely narrow, but they open out rapidly as the temperature falls. The $\alpha + \beta$ region is connected with both the Fe–Ni and the Fe–Al systems, and therefore forms a common bond between the 'open' and 'closed' types of diagram. At the melting point, the $\alpha + \beta$ region forms a peritectic system in connection with the Fe–Ni system. The $\alpha + \beta'$ region is a eutectic system, while

the $\alpha' + \beta'$ region forms another peritetic, to join up with the Ni–Al system. At the transition points, where the three-phase regions intersect the solidus and liquidus, four-phase reactions occur, as follows :

At 1380° C., liq. $+ \alpha' \rightarrow \alpha + \beta'$, at 4 Fe, 23 Al (at.-%)

At 1350° C., liq. $+ \beta \rightarrow \alpha + \beta'$, at 67·5 Fe, 12·5 Al (at.-%)

At 1340° C., there is a minimum in the liquidus and solidus at 50% Fe in the $\alpha + \beta'$ region.

Isothermals have been constructed, from 1350° C. downwards, at intervals of 100° C., to illustrate the progressive widening of the $\alpha + \beta$, $\alpha + \beta'$ and $\alpha' + \beta'$ gaps as the temperature falls. Photomicrographs are reproduced to illustrate the widening of the gap. They clearly show the characteristic two-phase patterns obtained after adequate heat-treatment at all temperatures from 1250° C. downwards, from alloys in the $\alpha + \beta$, $\alpha + \beta'$, and $\alpha' + \alpha'$ gaps. Later papers will illustrate microsections from the other two-phase regions, together with the corresponding three-phase triangles.

Contribution to the Morphology of Non-Metallic Inclusions in Steel Products—Inclusions in Ferrochromium. R. Zoja. (Metallurgia Italiana, 1949, vol. 41, Mar.–Apr., pp. 80–86). As a continuation of previous researches the inclusions in 96 specimens of ferrochromium were examined. Besides constituents already known and described, the presence of other characteristic constituents was noticed and observations were made on the frequency of the various types of inclusions (mainly crystallized chromium oxides, silicates, and sulphides). Two silicate constituents (α and β) formed 80% of the total inclusions observed, and other constituents (γ, δ, ϵ, ζ) derive from these by segregation. It is suggested that the constituents α and β are two conjugate solutions, the former rich in SiO_2, the latter in Cr_2O_3, and both containing other oxides and sulphides. Contrary to opinions previously stated, it was observed that a relation exists between the chemical composition of the iron alloy, especially its silicon content, and the type of the inclusions.

R. F. F.

An Investigation on Banding. J. D. Lavender and F. W. Jones. (Journal of The Iron and Steel Institute, 1949, vol. 163, Sept., pp. 14–17). Segregation in several steels has been studied by micro-radiographic methods. Temperatures of the order of 1200–1350° C. were required to remove banding, in agreement with rough calculations based upon diffusion data. This treatment is likely to result in overheating of the steel. The relation between the micro-radiographic and metallographic results has been considered.

Evaluation of Metallurgical Products by Means of Series of Microphotographs. E. von Graff. (Mikroskopie, 1947, vol. 2, pp. 30–41 : British Abstracts, BI, 1949, Jan., col. 64). Several series of photomicrographs are reproduced to illustrate varying degrees of sulphide-slag inclusion in rolled steel, grain-size growth, change of structure in cast iron as tempering proceeds with parallel change in physical properties, etc. It is suggested that similar series could be made to cover other variations in structure and properties.

Microradiography. R. Pospisil. (Non-Destructive Testing, 1949, vol. 7, Spring Issue, pp. 16–18). This is an English translation of a paper which appeared in Hutnické Listy, 1947, Jan., pp. 193–197 (*see* Journ. I. and S.I., 1947, vol. 157, Dec., p. 650).

New Precision Method of Measuring the Individual Lattices of Grains. Application to the Study of Cold-Work and Recrystallization. C. Crussard and F. Aubertin. (Société Française de Métallurgie, Oct. 20, 1948 : Revue de Métallurgie, Mémoires, 1949, vol. 46, June, pp. 354–359). The authors used the Bragg method of X-ray reflection spectroscopy with oscillating specimen, but by choosing an anticathode for each material examined such that the Bragg angle was about 88–90° (instead of the more usual 80–85°) an enhanced precision of the lattice measurements of 1/50,000, was obtained. In view of the great accuracy, the hollow specimen-holder was kept at a constant temperature by flowing water to avoid alterations to the lattice due to expansion. It was possible to measure the lattice constants of individual grains. The construction of the apparatus and the examination of the films are described, and the results obtained with cold-worked, high-porosity copper are discussed.—A. E. C.

The Rapid Determination of Orientations of Cubic Crystals. C. G. Dunn and W. W. Martin. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2637 ; Transactions of the American Institute of Mining and Metallurgical Engineers, 1949, vol. 185, pp. 417–427 ; Journal of Metals, 1949, vol. 1, July, Section 3, pp. 417–427). The orientation of cubic crystals is determined by comparing the pattern of Laue spots produced by X-ray diffraction with standard Laue photographs, and using isogonal and Leonhardt nets. A range of 75 standards has been prepared for comparison.—J. P. S.

Structure of Tempered Martensite. J. Mazur. (Nature, 1949, vol. 164, Aug. 6, pp. 230–231). From X-ray diffraction studies of steels of carbon content 0·89% and 1·20%, quenched in brine and immersed in liquid air for 1 hr., in liquid nitrogen for 30 min., and liquid helium at 1·5° K. for 6 hr., it is established that the decomposition of austenite into martensite is not completed at the temperature of liquid helium. When the specimens were tempered in boiling water at 100° C. for 1 hr., signs of the decomposition of martensite into ferrite mixed with precipitated cementite were evident from the X-ray diffraction spectra : this decomposition is faster in the 1·20% carbon steel than in the 0·89% carbon steel.—J. P. S.

Mechanism of the Formation of Pearlite in Steels. K. Honda. (Nature, 1949, vol. 164, Aug. 6, pp. 229–230). A theory for the lamellar nature of pearlite is brought forward, based on the mechanism of the A_1 transformation in a eutectoid steel. According to Honda, this takes place in two stages : austenite $\rightleftharpoons$ α-iron dissolving carbon $\rightleftharpoons$ pearlite. The first stage begins with many nuclei as centres, and involves a volume change of the order $\delta v/v = 8 \times 10^{-3}$, which sets up a pressure of the order of 30 kg./sq. mm. This is sufficient to cause slip bands in the α-iron grain, and carbon atoms in unstable conditions in the α-iron lattice leave the lattice and assemble in very thin cavities on the slip bands, resulting in lamellar pearlite. This explains why the pearlite lamellæ are parallel within the grain, but vary from grain to grain, and, since the concentration of carbon is equal on both sides of the slip-plane, why the lamellæ are of nearly constant thickness.—J. P. S.

The Course of the Isothermal Transformation of the Austenite of a Chromium Magnetic Steel, and the Influence of the Hardening Temperature and Hardening Time. W. Jellinghaus. (Mitteilungen aus dem Max-Planck-Institut für Eisenforschung : Archiv für das Eisenhüttenwesen, 1949, vol. 20, July–Aug., pp. 243–248). The isothermal transformation velocity of the austenite in a permanent magnet steel (C 0·97%, Si 0·30%, Mn 1·11%, and Cr 5·11%) between $+ 700°$ and $- 180°$ C. was measured. Using ordinary hardening temperatures there are three different temperature ranges with different characteristics. The pearlite stage is clearly separated from the intermediate stage by a zone of very low reaction velocity. As the temperature falls to about 380° C. the velocity of the intermediate stage transformation increases ; with further fall in temperature to the martensite point it gradually decreases. The transformation velocity of austenite quenched from 1050° is less than for that quenched from 850° C. and the intermediate stage transformation is almost completely suppressed. The high hardening temperature shifts the martensite point to lower temperatures. A long holding time at the hardening temperature delays transformation in the intermediate stage and in the upper part of the martensite stage. An attempt is made to explain intermediate stage transformation by the non-uniform distribution of the carbon in austenite, and by the growth of nuclei which form on the sudden transformation of the areas low in carbon.

R. A. R.

The Constitution of the Iron Alloys. E. Schwarz-Bergkampf. (Berg- und Hüttenmännische Monatshefte der Montanistischen Hochschule in Leoben, 1949, vol. 94, May, pp. 108–111). A scheme is proposed for the arrangement of the iron alloys according to numerical data on their limits of stability and miscibility. In this scheme the principal miscibilities, as given in Gürtler's handbook, can be read off from the line joining the melting points.—R. A. R.

CORROSION

Corrosion Testing. D. H. Garside. (Industrial Heating Engineer, 1949, vol. 11, Jan., pp. 21–23 ; Mar., pp. 41–44 ; May, pp. 96–99). Ordinary and accelerated corrosion tests are discussed with notes on the weatherometer, and Richard's

'industrial fog chamber.' The tentative British Standard Specification for a method of testing the performance of protective coatings is given and critically examined, and recommendations are made for conducting a series of tests. R. A. R.

The Atmospheric Corrosion of Ferrous Metals and Its Prevention. J. C. Hudson. (Year Book of the Coke Oven Managers' Association, 1949, pp. 116–144). *See* Journ. I. and S.I., 1948, vol. 160, Nov., p. 339.

Kure Beach Metal Corrosion Tests. E. C. Kreuzberg. (Steel, 1949, vol. 124, June 27, pp. 88, 91). A description is given of the marine corrosion station at Kure Beach, North Carolina, which is being used by an increasing number of metal producers and United States government departments. Among findings reported are that (1) in tidal conditions plain carbon steels deteriorate more rapidly than alloy steels, and more rapidly than they do in the atmosphere 800 ft. from the sea, (2) protective paints are more effective on alloy steels than on carbon steels, and (3) nickel plating offers better protection to carbon steel than does copper.—J. P. S.

Accuracy of Corrosion Tests with an Improved Apparatus. A. S. Afanas'ev, V. K. Rostovtseva, and M. G. Burakova. (Zavodskaya Laboratoriya, 1947, vol. 13, pp. 847–850 [in Russian]: Chemical Abstracts, 1949, vol. 43, Apr. 25, col. 2918). In a hermetic apparatus with thermo-regulator and ventilator, at constant humidity, loss-of-weight tests consisting of alternate immersion in water and drying in air at 35° ± 1°, with batches of 399, 59, and 201 steel sheet samples (first two batches sand-blasted, the third etched) were accurate within a mean relative error of respectively 0·74, 0·75, and 1·2%, mean relative square errors of 0·3, 3·5, and 3·9.

Note on Metallic Corrosion Research by Electron Diffraction. (Surface Structure of Acid-Resisting Alloys). S. Yamaguchi. (Scientific Papers of the Institute of Physical and Chemical Research, Tokyo, 1942, vol. 40, Oct., pp. 119–122). [In German]. The corrosion products on the surface of Hastelloy (Ni 60%, Mo 20%, Fe 20%) treated for 8–10 sec. with 7 *N* HCl at 80° C. were investigated by electron diffraction methods. They consisted of one oxide only, MoO_2.—R. A. R.

A Technique for Testing Corrosion Resistance under Adjustable Bending Stress. H. J. Seemann. (Metalloberfläche, 1949, vol. 3, Apr., Section A, pp. 85–86). A simple bend-test device is described in which a wedge-shaped specimen is clamped at one end; an adjustable load can be applied to the free end. The shape of the wedge is such that the stress is constant at all sections at increasing distances from the clamped end.—R. A. R.

Structure of Crystals of γ-$Fe_2O_3.H_2O$ Formed during Corrosion of Iron. N. A. Shishakov. (Journal of Physical Chemistry, U.S.S.R., 1948, vol. 22, Aug., pp. 953–955). [In Russian].

Corrosion Tests of Heater-Tube Materials for Acid Concentration. H. F. Brown. (Oil and Gas Journal, 1948, vol. 46, No. 52, pp. 134–137, 149: British Abstracts, BI, 1949, Jan., col. 51). A completely satisfactory material for heater thimbles in sulphuric acid concentrators has not yet been found. Initially a silicon–nickel alloy was used, which was later replaced in some cases by iron–silicon alloy or tantalum. Laboratory tests have been made on a wide range of materials. Cast 12–14% silicon–iron alloy has high acid resistance, but low impact strength and resistance to thermal shock. Iron–molybdenum–nickel alloy has inadequate corrosion resistance. Glass-coated carbon steel would be the most economical solution of the problem if satisfactory technique for handling and maintaining it could be established. Silver or silver coated tubes are unsatisfactory. Tantalum showed remarkable resistance to corrosion.

Measurement of the Corrosion of Welded Joints. I. I. Frumin. (Zavodskaya Laboratoriya, 1947, vol. 13, pp. 693–701 [in Russian]: Chemical Abstracts, 1949, vol. 43, Apr. 25, col. 2919). The depth of penetration of corrosion is determined by exploration with a profiiometer, preferably on plastic replicas. Intercrystalline corrosion is observed on specimens under stress. In the differential weight method the welded sample is exposed along with an unmoulded sample of the same metal and the same size. By this method copper steels welded with lowe-arbon steel proved to be less corrosion-resistant at the seam than in the mass. The contrary was found in welded manganese steels.

Electronographic Investigation of the Size of Iron Crystals and the Thickness of Oxide Films Formed on Their Surfaces. P. D. Dankov and N. A. Shishakov. (Journal of Physical Chemistry, U.S.S.R., 1948, vol. 22, Aug., pp. 956–960 [in Russian]: [Abstract] Metals Review, 1949, vol. 22, Feb., p. 27). This investigation was made by X-ray and electron microscopy. Data indicate that the film is composed at least of two elementary nuclei of gamma Fe_2O_3, and it is possible that one of them is Fe_3O_4 and the other Fe_2O_3. The range of thickness is about 16–18 Å.

Kinetic and Structural Factors Involved in Oxidation of Metals. E. A. Gulbransen. (Industrial and Engineering Chemistry, 1949, vol. 41, July, pp. 1385–1391). The author has studied the formation of oxide films on metals and oxidation reactions over wide ranges of temperature and pressure by measuring gains in weight with a sensitive quartz microbalance placed within a vacuum system. The original oxide film on a specimen of pure iron was removed by reduction with hydrogen and the gain in weight at room temperature when oxygen was added was plotted against time. A gain in weight equivalent to an oxide film 19 Å was noted. This film was stable to a vacuum of 10^{-5} mm. Hg and the thickness did not increase on adding more oxygen. Electron micrographs of films of Fe_3O_4 revealed a closely fitted mosaic structure for the oxide crystals; the thickness was not uniform and a second and third layer of crystals had formed over the first layer. Overlapping of crystals was noted at their crystal boundaries. This indicates that there is random orientation for crystal growth and that the direction of crystal growth varies from crystal to crystal.—R. A. R.

High Temperature Corrosion of Metals. A. Dravnieks and H. J. McDonald. (Corrosion, 1949, vol. 5, July, pp. 227–233). Rates of growth of scale on metals at high temperature may follow linear, parabolic, logarithmic, or asymptotic laws, most metals following a parabolic law. The theory most satisfactory in the explanation of this is due to Wagner where the metal–scale–gas system may be considered as a galvanic cell in which the scale is the electrolyte.—J. P. S.

Influence of Gas Adsorption on the Rate of Reaction between Metallic Oxides. H. Forestier and J. P. Kiehl. (Comptes Rendus, 1949, vol. 229, July 4, pp. 47–49). The authors have studied the influence of the adsorption of gases (water vapour, carbon dioxide, oxygen, air, nitrogen, neon, argon, helium) on the reaction of ferric oxide and nickel oxide when heated to various temperatures. The rate of formation of the compound Fe_2O_3NiO, at constant temperature and pressure, was an increasing function of the temperature of liquefaction of the gas used, hence of the amount of gas adsorbed. The general shapes of the production/temperature curves were similar, except that for CO_2 owing to the surface formation of $NiCO_3$ below 550° C.—A. E. C.

Report on the Conservation of Metals by Corrosion Control. G. Chaudron. (United Nations Scientific Conference on the Conservation and Utilization of Resources, Lake Success, 1949, Aug. 17–Sept. 6; Advance Copy). This is a general note on types of corrosion, of the corrosion resistance of certain metals and alloys, and on the impossibility of specifying corrosion protection without the fullest reference to surface preparation, useful life, and cost.—J. P. S.

The Rôle of Adsorption from Solution in Corrosion Inhibitor Action. N. Hackerman and H. R. Schmidt. (Corrosion, 1949, vol. 5, July, pp. 237–243). The extent to which inhibitors such as stearic acid are adsorbed on steel has been studied by a number of methods. In one a benzene solution of the inhibitor is shaken with a weighed amount of powdered steel and the content of the solution redetermined after equilibrium is reached. In a second, the inhibitor solution is run through a column of steel powder, the adsorption measured in a similar fashion, a corrosive liquid passed through and the extent of the corrosion measured, and finally, pure solvent is passed through and the degree of desorption ascertained. Only stearic acid indicates any measurable adsorption, and it is not completely desorbed. Electron diffraction methods confirm this, and that this is also true of a commercial inhibitor Kontol 115, the principal constituent of which is a ring compound containing a nitrogen atom and two carbonyl groups. J. P. S.

Present Knowledge of Low-Carbon 18-8. H. W. Gillett. (American Society for Testing Materials, 1949, Preprint No. 22). Data in the literature relating to stability of austenite,

precipitation of carbides, weldability, and corrosion-resistance of low-carbon 18/8 stainless steel are reviewed (46 references). To make this steel absolutely resistant to precipitation and grain-boundary corrosion, both carbon and nitrogen would have to be reduced to such low contents that it would be impracticable on an industrial scale. Commercially obtainable low-carbon and moderate nitrogen contents appear to be adequate to prevent the steel being susceptible to corrosion on cooling after welding. The test conditions must correspond to the service conditions for the correct evaluation of the susceptibility to corrosion.—R. A. R.

Corrosion of Turbine Journals. S. E. Bowrey. (Transactions of the Institute of Marine Engineers, 1949, vol. 61, Apr., pp. 57–71). Sea water may occasionally gain access to the turbine lubricating systems of ships and cause severe corrosion to the journals and gears. Corrosion-inhibiting turbine oils will not prevent corrosion by sea water, but sodium nitrite has been found to be successful as a sea-water corrosion inhibitor.—R. F. F.

Pickling, Etching, Pretreatment, Derusting, and Rust Inhibitors. R. Springer. (Metalloberfläche, 1948, vol. 2, May–June, pp. 123–130; Sept., pp. 199–203; Oct., pp. 224–229; Dec., pp. 271–275; 1949, vol. 3, Feb., pp. 45–50; Apr., pp. 96–99). The literature and patents relating to various forms of metal surface treatment as a preliminary to protective coating, for the period 1935 to 1943, are reviewed.—R. A. R.

Corrosion-Erosion of Boiler Feed Pumps. H. Seymour. (Cheap Steam, 1948, vol. 33, Sept., pp. 112–113: Fuel Abstracts, 1949, vol. 5, Mar., p. 82). Tests undertaken by the Detroit Edison Co. at Marysville (U.S.A.) power plant to find the most economical metal which can withstand attack by a particular feedwater are described. The feedwater used was condensate from a 75,000-kW. turbine generator containing approximately 1% of evaporated make-up. Total dissolved solids in the feedwater averaged less than 0·25 p.p.m. In selecting alloy material to be tested, 5% and 12% chromium steels, 18/8 alloy, 2 to 3% chromium steel, and a low-alloy chromium–nickel–molybdenum steel were included. The results indicated that 18/8 followed by 12% chromium steel were the most resistant materials.

Electrolytic Corrosion of Underground Pipes. L. T. Minchin. (Coke and Gas, 1949, vol. 11, May, pp. 153–156, 165). Tracing and preventing stray currents in underground gas mains are described with an example, and the four methods of preventing the corrosion employed for the gas mains in Belgium, where there is an extensive high-pressure grid system, are examined and their costs compared, the data being taken from a paper by R. de Brouwer (*see* Journ. I. and S.I., 1949, vol. 163, Dec., p. 477).

Crude Still Overhead System Corrosion. A. F. Blumer. (Corrosion, 1949, vol. 5, May, pp. 135–147). Corrosion problems at a crude oil distillation plant are considered. Chloride corrosion from salt in the crude oil tends to concentrate at the point where water first condenses. Removal of the salt compounds and H_2S before charging the crude oil to the distillation units is perhaps the most effective way of reducing corrosion. The acid constituents formed by the decomposition of salts, sulphur compounds, and carbonates can be effectively neutralized by contacting with alkaline solutions, *i.e.*, sodium carbonate for the crude oil charge and ammonia for final neutralization in the overhead stream. Recorders of the *p*H value are essential to proper control of the injection of ammonia.—R. A. R.

Corrosion Research in the Navy. J. J. Harwood and F. Schulman. (Corrosion, 1949, vol. 5, July, pp. 203–217) Corrosion research carried out under the Office of Naval Research, U.S. Department of the Navy, is described. The overall programme is grouped as follows: (1) Passivity and film formation studies. (2) Inhibitor and passivator studies. (3) Galvanic corrosion studies. (4) General corrosion studies. (5) High temperature oxidation and corrosion studies. Much of this work is being carried out in University departments.

Special Anodes for the Cathodic Protection of Water Tanks. A. L. Kimmel. (Corrosion, 1949, vol. 5, July, pp. 217–220). An oxide film may be artificially formed on iron and steel parts by making them cathodic to magnesium in fresh water. The film consists of magnesium hydroxide and is sufficient to protect the steel against further attack by this water, in the absence of the anode.—J. P. S.

On the Attack of Metals [Zinc, Cadmium, Nickel and Iron] by Moist Halogen Halide Vapours. W. Feitknecht. (Helvetia Chimica Acta, 1946, vol. 29, No. 7, pp. 1801–1815: Metallurgical Abstracts, 1949, vol. 16, May, p. 558).

ANALYSIS

The Determination of Small Quantities of Manganese in Steel with the Aid of the Steeloscope. B. N. Barkov and E. M. Rozhkov. (Zavodskaya Laboratoriya, 1947, vol. 13, pp. 184–186 [in Russian]: Chemical Abstracts, 1949, vol. 43, May 10, col. 3318). Two surfaces at right angles are ground on the sample and their common edge is given a radius of curvature of 0·5 to 1 mm. This edge is placed vertically, opposite a stationary copper electrode. A Sventitsky A.C. arc at 8 amp. is used for excitation. Determinations accurate to less than $\pm$ 20% are obtainable where manganese is between 0·02 and 0·15% by comparing manganese 4783·4 (*A*) with iron 4786·8 (*B*), 4779·4 (*C*), and 4788·7 (*D*). At 0·06% manganese $A = C$; at 0·08% manganese $A = D$; at 0·15% manganese $A = B$.

Sorting Thin Steel Wire with the Steeloscope. V. V. Borovkov and P. D. Korzh. (Zavodskaya Laboratoriya, 1949, vol. 15, June, pp. 744–746). [In Russian]. A steeloscopic technique for sorting steel wires is described and tables are given which are applicable to a wide range of steels. The method can be used with wire as thin as 0·02–0·03 mm.

Quantitative Spectrochemical Analysis of Cr–Ni (18/8) Steel, Using Solutions and a Hilger Medium Spectrograph. J. Eeckhout. (Analytica Chimica Acta, 1949, vol. 3, May, pp. 377–382). Operating conditions are described for the spectrochemical analysis of 18/8 steels using solutions and a spectrograph of moderate dispersion. The average deviation for one single exposure varied for the different elements from 4 to 9%.—R. A. R.

BOOK NOTICES

BLOM, A. V. "*Organic Coatings in Theory and Practice.*" La. 8vo, pp. xii + 298. New York, Amsterdam, London, 1949: Elsevier Publishing Co., Inc. (Price 32*s.*)

This book is an account of Dr. Blom's personal views on organic coatings, supported by numerous references taken mainly from American and Continental sources. The present position regarding our knowledge of coating materials is given in broad outline and suggestions are made for future research.

The book is divided into seven chapters. In the first, colloidal systems, the action of plasticizers and the processes of film formation are discussed. The next two chapters review the available film-forming materials from natural and synthetic sources respectively; there is much interesting and factual material in both of these chapters. Physical film formation by the evaporation of volatile ingredients or by 'congelation' of molten materials is discussed in the next chapter. This is followed by one on chemical film formation, in which air drying and stoving varnishes are considered. The succeeding chapter deals with the nature of pigments, their reaction with vehicles, and the dispersion of pigment particles. In the last chapter selected methods of testing are discussed.

The book covers a very wide field and consequently much has been omitted; this is fully admitted by the author, who states in his preface that he makes no claims for completeness. The use of paint for the protection of metals against corrosion is hardly mentioned; the book is, therefore, of limited interest to the metallurgist. In spite of its shortcomings, and many of these are due to the author's difficulty in expressing himself in clear and concise English, the book makes an important contribution to our knowledge of the subject and contains many stimulating ideas, which are valuable to all who are interested in the theory of surface coatings.—J. E. O. MAYNE.

CARMAN, P. C. "*Chemical Constitution and Properties of Engineering Materials.*" 8vo, pp. xii + 894. Illustrated. London, 1949: Edward Arnold and Co. (Price 50*s.*)

The title of this book is over-modest, as it does not suggest the extraordinarily wide field that the author

attempts to traverse. In the first place, 'chemical constitution' covers not only chemical analysis, but also molecular and crystal lattice structure, and secondly, the term 'engineering materials' has been interpreted with great liberality; the materials included, apart from metals and alloys, are ceramics, cements, glasses, natural and synthetic resins and rubbers, cellulose, proteins, polymerized substances, fuels, and natural waters. The aim of the book is to correlate properties of industrial significance with chemical composition and fine structure, and the subject is approached from a physico-chemical standpoint. The first chapters provide a résumé of modern theories of atomic and molecular structure, and the nature of the crystalline state (as revealed by X-ray analysis). Deformation of metallic crystals is described; and the nature of alloy systems and the significance of thermal equilibrium diagrams are dealt with at some length; a good many equilibrium diagrams of binary systems are reproduced in great detail. Thermal transformations in steel receive their due attention. The metals, ferrous and non-ferrous, are considered systematically, after which electrochemical corrosion and its prevention are discussed. Then follow chapters on all the other materials, with many lengthy excursions into branches of specialized chemistry, as, for example, the ternary lime–alumina–silica system, the theory of polymerization, and the chemical constitution of polysaccharides. The chapters on fuels and their combustion are a miniature treatise on fuel technology, and considerable space is given to flame propagation. Despite the wide variety of materials chosen by the author, there are still significant omissions. There seem to be, for example, only two passing references to lubricating oils, and only very brief mention of the nature, production, and uses of sintered carbides.

Each chapter concludes with references to book and recent periodical literature; 'classic' references have been deliberately omitted on the ground that the reader can always, if he wishes, work backwards from later articles.

Despite the fact that the book is crammed with useful and accurate information, one feels that the author has, perhaps, attempted too much, and it is difficult to see exactly what purpose it could best serve in a technical library. It contains too much detail to be a satisfactory elementary introduction to industrial physical chemistry, particularly for an engineer, and, as the author himself clearly appreciates, it cannot take the place of standard reference books in the various branches of the subject. Its best function would appear to be that for which the author himself used it, namely, as the basis of a lecture course in industrial chemistry—with the proviso that there are some important omissions which should be made good.

Finally, the book is most attractively produced, and paper, printing, and illustrations are all of excellent quality.—M. A. Vernon.

Conservatoire National des Arts et Metiers. "*Semaine d'Etudes de la Physique des Metaux. (Juin* 1948)." Comptes Rendus publiés sous les auspices de l'Institut de Recherches de la Sidérurgie (IRSID). 4to, pp. 99. Illustrated. Paris: Centre de Documentation Sidérurgique.

This publication contains the following papers presented at a symposium on the physics of metals held in Paris in June 1948. Abstracts of the articles of ferrous interest will appear in forthcoming issues of the *Journal:*

Chevenard, P.: Opening address.

Laurent, P.: "Nucleation Phenomena in Transformations of Steels."

Sourdillon, A.: "The Decomposition of Austenite in the Isothermal Condition and the Practical Utilization of this Reaction for the Treatment of Steels."

Crussard, C.: "Contribution to the Theory of Martensitic Transformations."

Michel, A.: "Martensitic Hardening and Tempering."

Jolivet, H.: "General Note on Austenite Decomposition."

Lacombe, P.: "Hardening and Tempering Light Alloys."

Herenguel, J.: "The Industrial Aspect of the Hardening and Annealing of Light Alloys."

Dietz, A. G. H. "*Engineering Laminates.*" 8vo, pp. viii + 797. Illustrated. New York, 1949: John Wiley and Sons, Inc.; London: Chapman and Hall, Ltd. (Price 60*s.*)

It is well known that laminated materials are not a new idea, but one reader of Dr. Dietz's book, at least, was surprised to learn that there exist fragments of laminated wood dating from the third Egyptian dynasty (*circa* 2780 B.C.) and of hot-hammered silver-clad copper, also Egyptian, even earlier. The use of laminated material in engineering construction has become more extensive in recent years, with the production of new materials, and there has been a need for a comprehensive reference book such as the present one. It takes the form of a symposium of 22 chapters, each written by an author or authors who are expert in their particular subject, with Dr. Dietz as editor. The individual authors have not been hampered by having to follow a standard pattern in writing their contributions. The scope of the book is defined in the preface as covering laminates either made of a single material, or composite, which are "useful from an engineering standpoint" and "significantly different in their composite properties or use from their constituents." Laminates which are purely decorative, and those in which the surface layer is purely protective, are specifically excluded; thus, for instance, tinplate and galvanized iron are excluded, but there is a complete section on sprayed metal. The first section, the longest in the book, is a remarkably complete mathematical treatise on the strength of laminates and sandwich structural elements. The next section, on adhesives, is written rather from the chemical standpoint, but the latter part deals adequately with properties and methods of testing. Then follow a series of contributions on the different types of laminates; these, in general, describe their manufacture, properties as affecting their uses, and applications. All important laminates, metallic and non-metallic, appear to be included; those of particular metallurgical interest are thermostat metals, aluminium-clad products, hot-dipped aluminium-coated steel, copper and copper-alloy clads, nickel and nickel-alloy clad steel, stainless clad steel, cast laminated metallic materials, hard-surfacing by fusion welding, sprayed metal, and glass-lined steel equipment. It is quite impossible to give a short adequate summary of the considerable mass of information, qualitative and quantitative, in these chapters, but as an example, the chapter (a short one) on stainless clad steel may be cited. This gives a brief historical introduction, followed by a summary of methods of manufacture—casting, fusion melting, intermelting, and the sandwich method (rolling). Compositions of A.I.S.I. (American Iron and Steel Institute) steels are given, and corrosion resistance, annealing and pickling, mechanical properties, fabrication, and uses are discussed. This particular chapter, as has been pointed out, is one of the shorter ones; others, particularly that on hard-surfacing (which also discusses hardness-testing and wear resistance) contain substantially more information.

The text is fully annotated throughout with bibliographical references, including American patent literature; and the diagrams and production generally are of a very high standard. It should prove an invaluable work of reference to engineers and others concerned with the production and use of this type of material.—M. A. Vernon.

F.I.A.T. Review of German Science, 1939–1946. "*Inorganic Chemistry.*" Parts IV and V. Senior Author: Wilhelm Klemm. 8vo, pp. 260 + 315. Berlin, 1948; Office of Military Government for Germany, Field Information Agencies, Technical.

These two volumes of the six-part series on Inorganic Chemistry (*see* notices in Journ. I. and S.I., 1949, vol. 161, p. 279 and vol. 162, p. 493) are devoted to General Researches, whereas their predecessors dealt with Special Inorganic Chemistry. Part IV deals first with the constitution of inorganic substances, their molecular and ionic properties, and includes an interesting and valuable section on the crystal structure of intermetallic phases where, in addition to lattice constants, the types, in similarity with better known compounds, are given for a number of less important ones. Other properties of inorganic compounds which are dealt with include X-ray emission and absorption spectra, and magnetic, electrical and optical phenomena. The conclusion of this part covers thermo- and electro-

chemistry ; in the former appear vapour-pressures, specific heats, entropy values, and so on ; in the latter, dielectric constants, electromotive force and the data of electrolysis. Part V opens with a long section on equilibrium in both vapour and liquid phases, in which the passages on, for example, the equilibrium between sulphur and the sulphides of various metals, between molten alloys and the vapours of their constituents, and on the solubility of complex salts, will be of direct interest to the metallurgist. This part concludes with some hundred pages on the scientific basis of metallurgical processes : the deoxidation of steel, the treatment of vanadium-containing slags, the chemistry of the blast-furnace process, the desulphurization of pig iron with soda ash, and the reactions of molten iron with sulphide slags, with sulphide-silicate slags, and with gases, are dealt with from the theoretical point of view. A similar treatment is given to processes for the extraction of other heavy metals such as copper, lead, and zinc, and to that of aluminium, where considerable space is devoted to work on the preparation of alumina from bauxite, and on both electrolytic and non-electrolytic methods for the extraction of aluminium. By comparison, the section on magnesium with which this volume concludes, is brief. The standards of production of this series remain high.—J.P.S.

GLOCKER, R. "*Materialprüfung mit Röntgenstrahlen unter besonderer Berücksichtigung der Röntgenmetallkunde.*" Dritte Auflage. 8vo, pp. viii + 440. Illustrated. Berlin, Göttingen, Heidelberg, 1949 : Springer-Verlag. (Price DM. 58.)

Some readers may be familiar with the first two editions of this work, although the second appeared during the war years. It is avowedly written from the point of view of the physical metallurgist, and appears to be intended primarily as a guide and reference book for the practical worker. Thus, for example, formulæ are quoted, and not derived : this, indeed, would scarcely be possible in so compact a treatise. In this, the third edition, the section on the examination of coarse structure has been expanded, and that on X-ray spectral analysis abridged ; the chapters on broadening of lines and measurement of strain have been rewritten ; and there is a new chapter dealing briefly with the distribution of atoms in non-crystalline materials. The remaining chapters have been brought up-to-date.

Professor Glocker develops his subject conveniently and logically ; after the briefest of introductions on the characteristics of X-rays, and their uses for testing materials, he proceeds at once to a descriptive account of modern X-ray tubes, and complete generating installations, including the necessary protective arrangements. The properties of X-rays are next summarized—absorption, secondary radiation, reflection, diffraction, etc.—and their detection and measurement by photographic and ionization methods. The three main sections of the book follow, dealing respectively with ' coarse structure ' examination (*e.g.*, detection of defects), X-ray spectral analysis, both qualitative and quantitative, and ' fine structure ' determination, *i.e.*, X-ray crystallography. The latter section, which is preceded by an introductory chapter on basic crystallographic principles, is by far the most extensive, comprising, as it does, two-thirds of the book. This is not entirely due to the greater complexity of this branch of the subject ; the author is probably influenced by his appreciation of the increasing potentialities of X-ray crystallography as a research tool. All three sections are dealt with comprehensively and, within the limitations imposed by space, adequately ; and they are reasonably up-to-date ; for example, the possibilities arising out of the use of ultra-hard X-rays produced with the aid of particle accelerators such as the betatron are discussed. Two chapters may be of particular interest to physical metallurgists ; one deals with the crystal structure of alloys and the construction of phase diagrams, including the iron–carbon system, by X-ray methods ; and the other gives an excellent account of the use of X-rays in measuring elastic strain. There is a mathematical appendix to the third section, and the book concludes with a selected and classified bibliography. (There are hardly any bibliographical annotations to the text). The book is well produced : line diagrams are excellent throughout, and photographs reasonably clear, although the photographs accompanying descriptions of equipment serve no useful purpose other than giving the reader some idea of its external appearance.

The book should be chiefly useful to the student or investigator who finds it necessary to obtain a general knowledge of the techniques with which it deals ; he will find rather more detail than is customary in books of similar scope ; but he will certainly need to read more extensively if he wishes to work himself in the same field. Even in the latter case, however, the present work might at least serve as a useful introduction.—M. A. VERNON.

LOW, B. B. "*Strength of Materials.*" 8vo, pp. vii + 231. Illustrated. London, 1949 : Longmans, Green and Co. (Price 12*s.* 6*d.*)

Few students find ' Strength of Materials ' an easy part of their curriculum, and most may well feel apprehensive when their first introduction to the subject takes the form of one of the standard textbooks, most of which are of an impressive length and complexity. The author of this work, an experienced teacher, is well aware of the difficulties encountered, and has attempted the task of compressing the material into a smaller span, and at the same time simplifying it as far as possible. Whether this is really practicable, or whether, in the long run, it will make things easier is a moot point, and the proof of this particular pudding will be in the eating. In furtherance of his aim, the author has also departed from the strictly logical order of presentation, in order that the student may first tackle the simpler, and more easily applied parts of his subject. Proofs of theorems, although complete, are inevitably somewhat compressed, and they may offer some temptation to be learnt parrot-wise. Great care has obviously been taken with the diagrams, which are excellent, and another most helpful feature of the book is the inclusion of worked examples in the text, illustrating the application of the various theorems. A lavish selection of exercises, some of which are questions taken from University and Institution examination papers, concludes each chapter, and answers are given at the end of the book. If the student is able satisfactorily to work through these, he may well be considered to have learnt his subject. The book will merit the attention of teachers who feel the need for a primer for students of engineering or physics who have the necessary mathematical knowledge.—M. A. VERNON.

PLANCK, MAX. "*Scientific Autobiography and Other Papers.*" With a Memorial Address by Max von Laue. Translated by Frank Gaynor. 8vo, pp. 192. New York, 1949 : Philosophical Library. (Price $3.75)

Although this book contains nothing on metallurgy, the quantum theory has so profoundly influenced the theoretical treatment of the properties of metals that this small volume calls for notice here. Max Planck lived to be nearly 90. In the Second World War his house was destroyed and his fine library disappeared. He was buried for some hours in the ruins of an air-raid shelter, and was then given asylum by the Americans.

The short section called an autobiography gives only the bare record of his academic career and notes on his teachers, with a very modest account of his own contributions to science. The essays are entirely on the philosophical aspects of science. There is much in his views, with their distrust of metaphysical ideas, that recalls the writings of Mach, Ostwald, and Karl Pearson. He is always lucid, and has been well served by his translator. The principle of uncertainty might seem to some to have destroyed the foundations of physical science ; that this is not so is clearly shown. All who knew his work and had heard him speak, if only once, must have realized that he was a great man.—C. H. DESCH.

REINER, M. "*Deformation and Flow.*" An Elementary Introduction to Theoretical Rheology. 8vo, pp. xix + 346. London, 1949 : H. K. Lewis and Co. (Price 32*s.* 6*d.*)

Only a small part of this work, emanating from Palestine, refers to metals. Liquids, cements, and such substances as rubber, are fully discussed. The very short section on the work-hardening of metals includes an account of experiments by the author in which tensile specimens were elongated and then turned down to the smallest section produced during the elongation, this operation being repeated several times. The changes produced in the process are discussed mathematically.—C. H. DESCH.

NEW PUBLICATIONS

RICHARD THOMAS AND BALDWINS LTD. "*R.T.B. Methods of Analysis.*" Large 4to, pp. 112. London, 1949: published by the Company. (Price 42*s.*)

The purpose of this volume is to place on record, for their own laboratories, the methods used by each section of the R.T.B. organization, and it is a co-operative work by their leading metallurgists and chemists. The scope of the company's activities necessitates the analysis of steels, cast irons, ferro-alloys, ores, slags, fluxes, refractories, fuels, pickling solutions, and water, on the ferrous side, and brasses, bronzes, white metals, ingot-tin, and aluminium alloys on the non-ferrous side. The methods employed are all well-established, and gravimetric or volumetric in nature, the only exceptions being a colorimetric estimation of small quantities of iron in acid refractories, the use of the Spekker absorptiometer for the colorimetric analysis of aluminium alloys, and the use of ferri-cyanide paper in the estimation of the porosity of tinplate. Useful tables, of atomic weights, densities, and melting points, of the equivalents of standard and miscellaneous solutions, of the strength of normal solutions, and of selected gravimetric factors are included. The volume is very well produced, but the price seems excessive in comparison with standard textbooks on analysis.—J. P. S.

VEREIN DEUTSCHER EISENHÜTTENLEUTE. "*Gemeinfassliche Darstellung des Eisenhüttenwesens.*" 15 Auflage. 8vo, pp. vi + 258. Illustrated. Düsseldorf, 1949: Verlag Stahleisen m.b.H. (Price 14.50 DM.)

There has been a lapse of twelve years since the publication of the fourteenth edition of this work. During this period, technical developments have been rapid, and the task of bringing this fifteenth edition up-to-date has been extensive. In the main it has been carried out adequately, but there are one or two surprising omissions, in fields in which German technicians have met with notable success. The general scheme follows conventional lines—historical introduction, the blast-furnace process, steel melting, shaping, properties and tests, works layout and equipment, and finally, an economic appendix. Naturally, the publishers have German practice in mind throughout the book, but this does not detract from its interest or value. The historical introduction contains a chronological table (which, however, stops short at 1919). The raw materials section is notable for its excellent treatment of ore beneficiation, and the Krupp-Renn and Lurgi processes are not forgotten. In the chapter on blast-furnaces, blast enrichment is dismissed very briefly, but the acid smelting process is described. Considerable attention is given to electric blast-furnaces. The Bessemer process is described in great detail, but there is no mention either under "Steel Production" or in the following chapter on casting, of the side-blown converter. Apart from this, the descriptions of iron and steel founding are remarkably complete. Plastic deformation is dealt with adequately, but more space might have been given to the concluding part of this chapter, which deals with surface protection very briefly indeed. There is, for instance, no mention of phosphate coatings. Practical heat-treatment is not described, although thermal transformations and the iron–carbon diagram are considered at some length under "Properties and Tests." Hardenability testing is not mentioned; neither are **S**-curves and their uses described. The technical part concludes with a selected bibliography of works in the German language on iron and steel manufacture; the books quoted cover some of the points the omission of which has been noted. The second part of the book contains production statistics, as far as they are available, for the chief producing countries of the world, over a number of years.

In general, the book is lucidly written, profusely illustrated with line drawings and photographs, and contains much useful quantitative data.—M. A. VERNON.

NEW PUBLICATIONS

AITCHISON, L. "*Introduction to Industrial Metallurgy.*" 8vo, pp. xii + 456. London, 1949: Macdonald and Evans. (Price 30*s.*)

AMERICAN IRON AND STEEL INSTITUTE. "*Annual Statistical Report* 1948." 8vo, pp. 196. New York, 1949: The Institute.

AMERICAN IRON AND STEEL INSTITUTE. CONTRIBUTIONS TO THE METALLURGY OF STEEL No. 35. "*Survey of Open Hearth Alloy Steel Off-Heats and Residual Elements.*" 4to, pp. 23. New York, 1949: The Institute.

BARKSDALE, J. "*Titanium. Its Occurrence, Chemistry and Technology.*" Pp. 570. New York: The Ronald Press Co. (Price $10)

BLAIR, G. W. SCOTT. "*A Survey of General and Applied Rheology.*" 2nd edition. Pp. xx + 314. Illustrated. London, 1949: Sir Isaac Pitman and Sons, Ltd. (Price 40*s.*)

BRITISH INTELLIGENCE OBJECTIVES SUB-COMMITTEE, Overall Report No. 15. "*The Ferrous Metal Industry in Germany during the Period* 1939–1945." By Geo. Patchin and E. Brewin. 8vo, pp. 270. London, 1949: H.M. Stationery Office. (Price 4*s.* 6*d.*)

BRITISH IRON AND STEEL FEDERATION. "*Statistics of the Iron and Steel Industries of Overseas Countries for* 1947." (Statistical Year-Book for 1947. Part II.) 8vo, pp. xiv + 137. London: The Federation. (Price 15*s.*)

BRITISH STANDARDS INSTITUTION. B.S. 1563: 1949. "*Cast Iron Sectional Tanks (Rectangular).*" 8vo, pp. 34. London: The Institution. (Price 5*s.*)

BRITISH STEEL FOUNDERS' PRODUCTIVITY TEAM. "*Steel Founding.*" Report of a visit in 1949 of a 'Productivity Team' representing the Steel Founding Industry. 4to, pp. x + 108. Illustrated. London, 1949: Anglo-American Council on Productivity. (Price 3*s.*)

BRÜHEIM, G. "*Schleifen.*" 4. neu. bearb. Aufl. (Folge 7 der "Werkstattkniffe"). Pp. 75. Illustrated. Munich, 1949: Carl Hanser-Verlag. (Price DM. 3.60)

CHALMERS, T. W. "*Historic Researches.*" Chapters in the History of Physical and Chemical Discovery. 4to, pp. 287. Illustrated. London, 1949: Morgan Brothers (Publishers), Ltd.

DEN HARTOG, J. P. "*Strength of Materials.*" New York: McGraw-Hill Book Co., Inc.; London: McGraw-Hill Publishing Co., Ltd. (Price $4.)

DEUTSCHER NORMENAUSCHUSS. "*Werkstoffnormen Stahl, Eisen, Nichteisen-Metalle.*" DIN-Taschenbuch 4, 17 Aufl. Pp. 271. Berlin and Krefeld-Uerdingen 1949: Beuth-Vertrieb-G.m.b.H. (Price DM. 17)

"*Engineering Materials. A Textbook and Work of Reference for Students and Civil Engineers.*" London: Blackie and Son, Ltd. (Price 17*s.* 6*d.*)

IMPERIAL INSTITUTE. "*The Mineral Industry of the British Commonwealth and Foreign Countries. Statistical Summary. (Production, Imports and Exports)* 1941–1947." 8vo, pp. 339. London, 1949: H.M. Stationery Office. (Price 21*s.*)

JACOBSON, C. A. "*Encyclopedia of Chemical Reactions.*" Vol. III, pp. xi + 842. New York: Reinhold Publishing Corp. (Price $12)

MILLER, A. R. "*The Adsorption of Gases on Solids.*" 8vo, pp. 100. London: Cambridge University Press. (Price 10*s.* 6*d.*)

PATTY, F. A. "*Industrial Hygiene and Toxicology.*" Volume II, 8vo, pp. xxviii + 535–1138. New York, 1949: Interscience Publishers Inc. (Price 90*s.*)

RYSCHKEWITSCH, E. "*Oxydkeramik des Einstoffsystems vom Standpunkt der physikalischen Chemie.*" 8vo, pp. vi + 280. Illustrated. Berlin, Göttingen and Heidelberg, 1948: Springer-Verlag. (Price DM. 36)

SPREADBURY, F. G. "*Permanent Magnets.*" London: Sir Isaac Pitman and Sons, Ltd. (Price 35*s.*)

STEVENS INSTITUTE OF TECHNOLOGY. "*Determination of the Effect of Particle Size on the Properties of Commercial Iron Powders and Compacts Made from these Powders by Conventional Cold-Pressing and Sintering Techniques.*" Pp. 306. Washington: U.S. Department of Commerce, Office of Technical Services.

WARNOCK, F. V. "*Strength of Materials.*" Seventh edition. London: Sir Isaac Pitman and Sons, Ltd. (Price 15*s.*)

ZEYEN, K. L. "*Neue Erkenntnisse und Entwicklungen beim Schweissen von Eisenwerkstoffen.*" 8vo, pp. 214. Illustrated. Munich, 1949: Carl Hanser Verlag. (Prices DM. 10.80 unbound; DM. 12.80 bound.)

WILLIAM BARR, A.R.T.C., F.I.M.

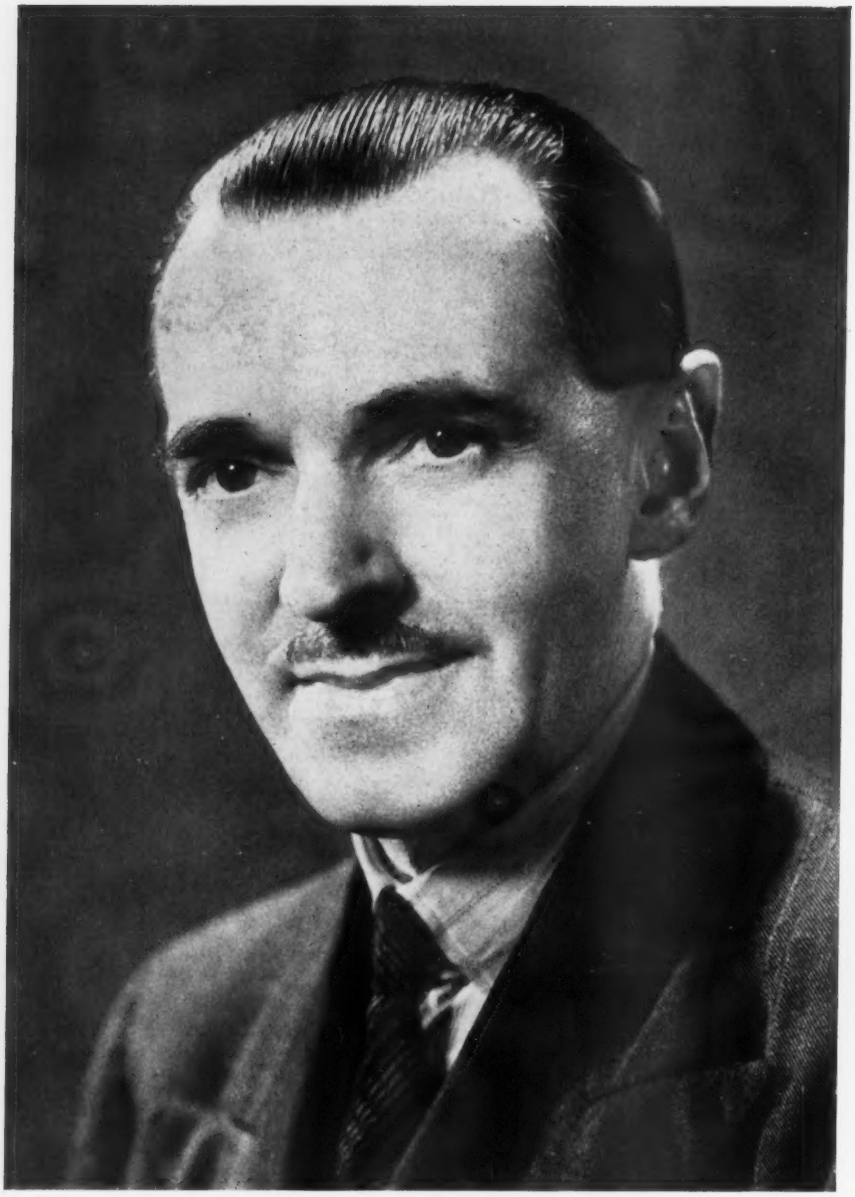
WILLIAM BARR, Chief Metallurgist and Executive Director of Colvilles, Ltd., and a Director of The Fullwood Foundry Company, Ltd., was born at Larkhall, Lanarkshire. He was educated at the Hamilton Academy and at the Royal Technical College, Glasgow, where he gained the Diploma in Metallurgy in 1921, and the Associateship of the College in the following year.

In the First World War, Mr. Barr served in the Special Brigade, Royal Engineers, and was mentioned in despatches. After completing his academic training he joined the staff of Messrs. David Colville and Sons, Ltd., in 1922, as a metallurgist. In the following year he was appointed Assistant Works Metallurgist at the Dalzell Works. After the formation of Colvilles, Ltd., he became Chief Metallurgist of the company in 1936. In 1945 Mr. Barr was appointed a Director of The Fullwood Foundry Company, Ltd., and in 1947 he became Executive Director of Colvilles, Ltd.

During his long association with Colvilles, Mr. Barr has contributed articles and technical papers to numerous societies, his activities being specially devoted to researches on creep, the brittle fracture of mild steel, and the development of low-alloy high-tensile structural steels and stainless-clad steels. In the Second World War he was a pioneer in the development and production of armour and bullet-proof plates by the Basic Open-Hearth Process. He was a member of the British Mission which visited the U.S.A. in 1943 in connection with the conservation of alloys in steel production.

Mr. Barr's specialized knowledge of alloy steels and welding has been recognized by his service on many technical committees : the Technical Advisory Committee of the Special and Alloy Steels Committee ; the Tank and Aircraft Armour Advisory Committee ; the Admiralty Armour Technical Committee (he is Chairman of the Sub-Committee) ; the Admiralty Ship Welding Committee ; the Ferrous Metals Committee of the Inter-Service Metallurgical Research Council, and the J.E Committee of the British Electrical and Allied Industries Research Association (of which he is Chairman of Sub-Committee J.Ec.). He is also a member of the Metallurgy Panel of the British Iron and Steel Research Association, and is Chairman in Scotland of the Joint Committee for National Certificates in Metallurgy.

In 1944 Mr. Barr was asked to take part in the founding and organizing of the Institution of Metallurgists, of which body he is a Founder Fellow and a Member of Council. He has been a Member of Council of the British Welding Research Association since 1947, and of the Iron and Steel Institute since 1948. In 1949 he was elected President of the West of Scotland Iron and Steel Institute. Mr. Barr became a Member of The Iron and Steel Institute in 1925.



William Barr. A.R.T.C., F.I.M.

Member of Council